metal treatment and Drop Forging

Vol. 28 : No. 184 JANUARY, 1961 Price 2/6

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for
Maximum Production and Precision

Simple to operate Massey Mechanical Forging Presses are designed for the mass production of precision drop forgings. They can be used for three distinct purposes - die forging, extrusion and coining.

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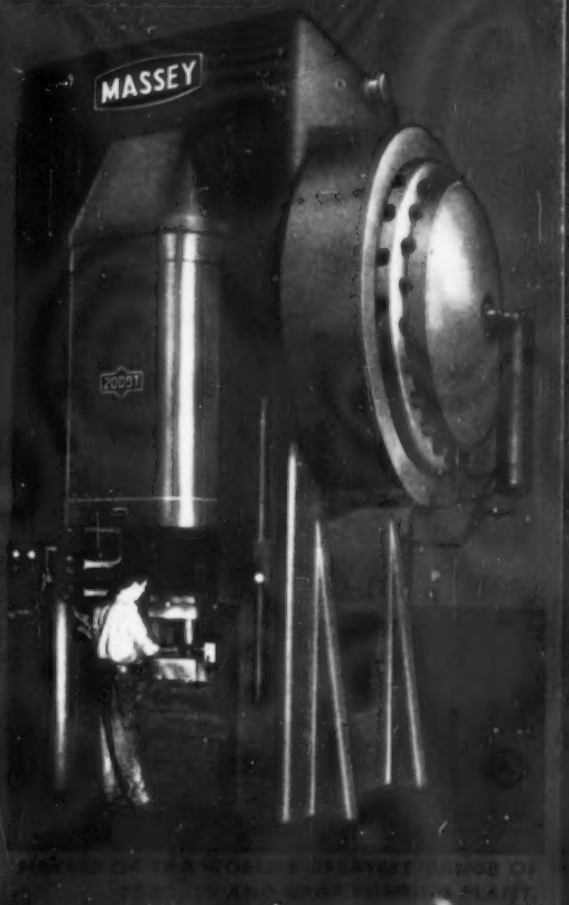

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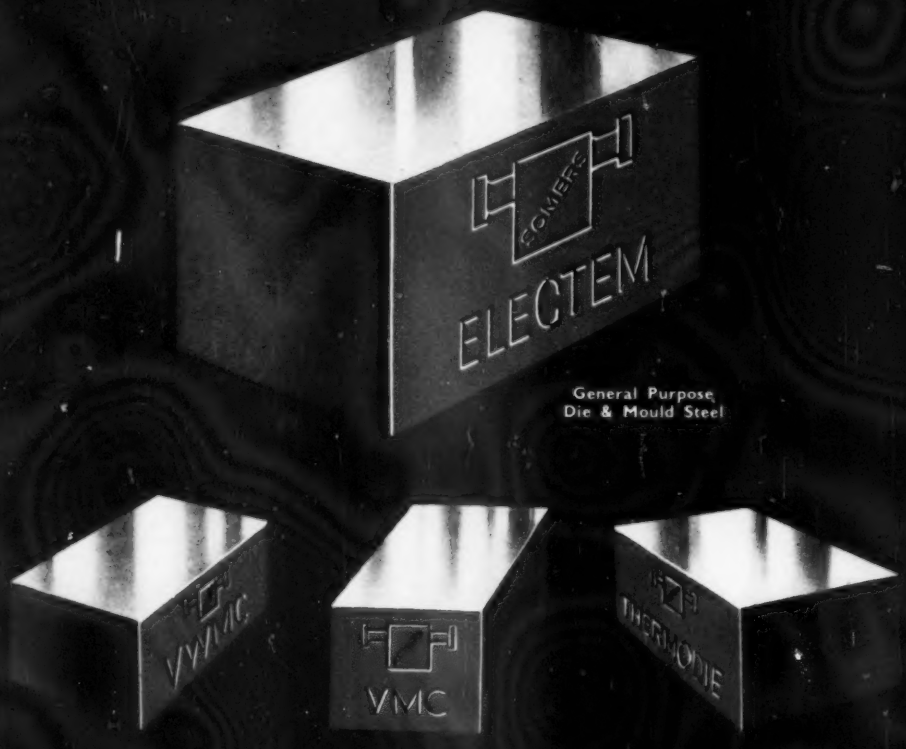

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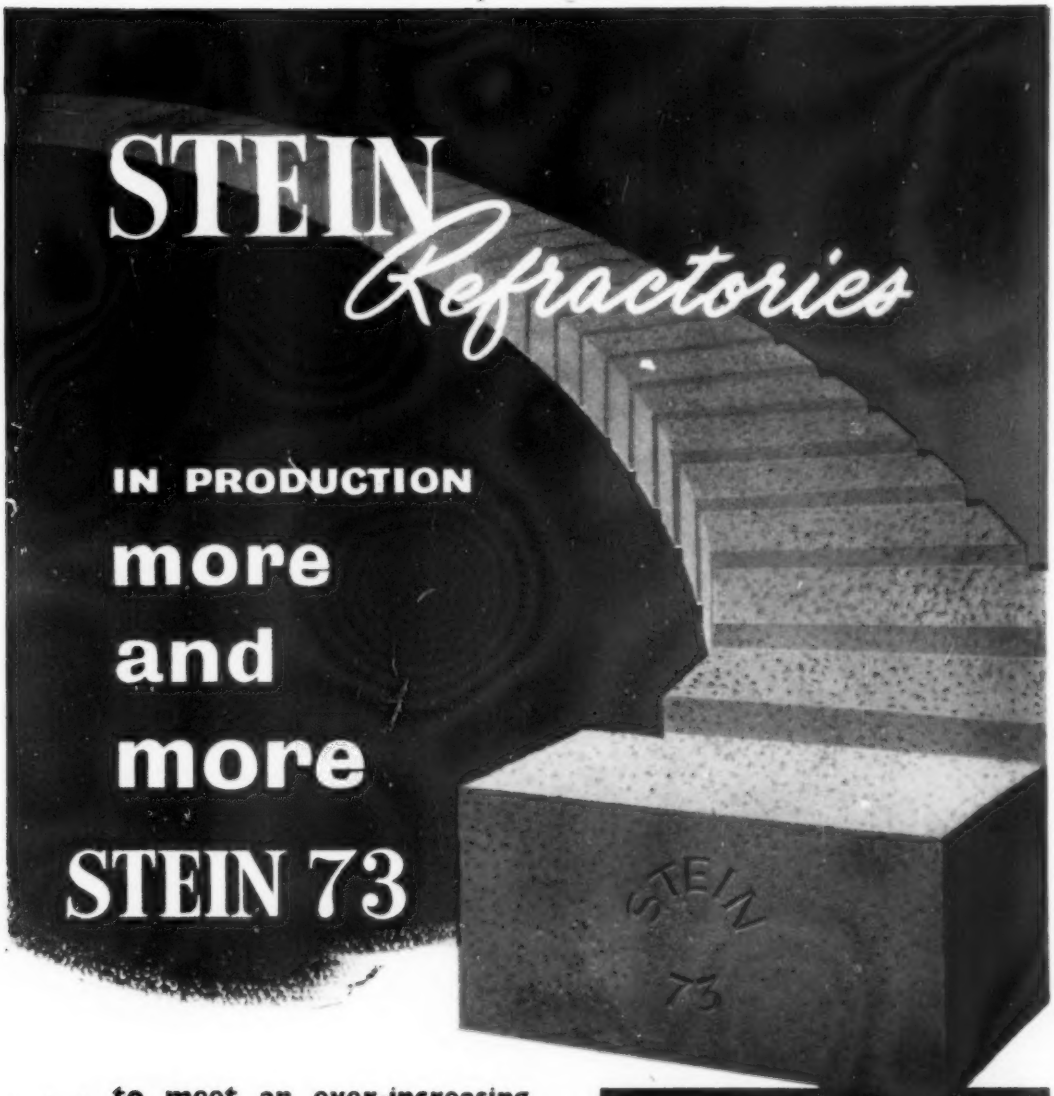

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Refractoriness — under load 28 lb./in.2 (2kg./cm.2) — 5% deformation at	1660°C
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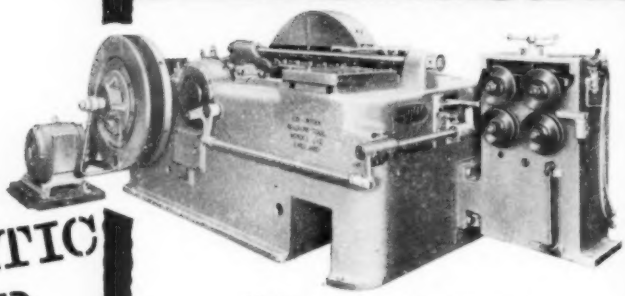

No. 2 machine with guards removed

The stock is placed between the rolls and is then fed through the dies to the workstop. The die then closes, cutting off the stock to the required length, and delivers it to the heading position. The heading tool then completes the forging, which falls down the chute and is taken away by conveyor.

The Automatic Forger can be supplied in several sizes and for making balls of up to and including 4-in. diameter.

No. 3 machine illustrated

These machines are made in three sizes. The smallest handles bars up to 2-in. diameter. The largest size, as illustrated, takes bars up to 4-in. diameter. They are an invaluable asset to any forging or drop stamping plant.

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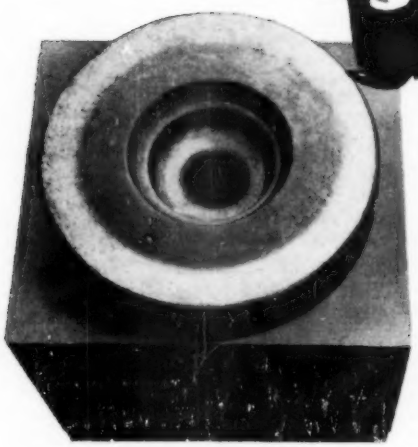

"4,000 ops and still going strong —

(It's all that lubrication with a FOLIAC Colloidal Graphite Dispersion)"

Foliac Colloidal Graphite supersedes the traditional methods of die lubrication. It not only eliminates metal-to-metal contact, but protects the surfaces of the dies from corrosion, cracking and burning. It provides a tough, self-lubricating film over the entire die surface that reduces friction and improves metal flow. The result is a consistently well finished product from a die that lasts twice as long—or longer. Why not ask our representative to call to discuss specific problems, and advise on applications.

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A1

STEEL-A RECORD YEAR

Biggest-ever output flows from Britain's busy furnaces

BRITAIN'S steel production in 1960 was the greatest ever recorded: over 24 million ingot tons. Output was running close to capacity in almost every section of the industry. Total production was *twenty per cent* above 1959.

Britain's capacity to produce steel has been increasing by leaps and bounds as a result of the £900 million which has been spent on three great development programmes. Last year alone the steel industry spent £130 million on development, *one-sixth* of all the money spent by manufacturing industry on capital investment in this country.

In the next four years the industry has already planned to spend another £450 million, and to raise its production potential by nearly one-third.

Another article on the opposite page describes a great new development in Scotland, which brings with it the promise of greater prosperity across the Border. Also in hand is a vast project near Newport, Monmouthshire. This is the Spencer works of Richard Thomas & Baldwins, which when completed will be Europe's most modern integrated steelworks. An enormous green-field site has been laid out, and work is going ahead at a tremendous pace.

Expansion under way

Three more new mills to roll improved and more efficient sections for building and constructional work will come into operation shortly at Colvilles, South Durham and the United Steel Companies.

Among the very large schemes announced in 1960 were the Tinsley Park scheme of the English Steel Corporation to increase their output of alloy and special steels, costing £26 million; the United Steel Companies' plans at Appleby-Frodingham and Samuel Fox, costing £32½ million; a Dorman Long plan costing £36 million; and the entirely new works at Rotherham of the Park Gate Iron and Steel Company which will cost £55-60 million.

Prospects for '61

What are the prospects for 1961? The industry will be able to improve on the 1960 record figure. Capacity in 1960 was nearly 26 million tons – this year it will be 27 million tons. All the time, the industry is expanding to meet future demand. This expansion is geared, not to any yearly fluctuations in our manufacturing activity, but to the steadily rising curve. It is in the confidence that this curve *will* go on steadily rising, that the many far-reaching development plans in steel are being pressed forward without slackening.

Exports a record too

Besides the enormous amount of steel which Britain exported last year in the form of manufactured goods, the steel industry's own direct exports of steel were booming.

Over 4 million ingot tons, valued at some £230 million, went into direct exports of steel in 1960 – a post-war record.

Latest available figures show that well over one-third of this went to Commonwealth countries: exports to the Commonwealth were 20% up on 1959 in tonnage. Tonnage of exports to Western Europe showed a similar upsurge. Exports to the Soviet Bloc including China, though small by comparison with total exports, also showed an increase.

Two new steels for supersonic flight

What will the supersonic airliner of the future be made from? Opinions are divided, but the Bristol T.188 research plane, which has been built to explore the problem of flight at speeds of the order of 1,800 m.p.h., is made from heat-resistant stainless steel to withstand the high temperatures caused by the friction of the air at such speeds.

The steel scientists of Sheffield have produced two new types of steel from the experience of the Bristol T.188. Although heat-resisting alloy steels had already been in use in jet engines for many years, these were not suitable for making large thin sheets of the kind required to build an airframe. The new steels produced in Sheffield are capable of being made in large sheets, and can be "worked" into the complicated shapes required.

New bridges hang on steel

The new road bridge over the Firth of Forth is being anchored into the solid rock with specially thick steel cables. They are 1¾ in. in diameter and have a breaking load of more than 350 tons. Over a thousand of these cables, enclosed in groups of four in 4½ in. diameter steel tubes, are embedded 240 feet into the rock. Other similar cables are being used to support the 110 ft. high towers until the main steel suspension cables are in position.

Other new bridges using steel suspension cables are planned over the Wye and Severn. Altogether some 70,000 tons of steel will go into the three bridges.

STEEL–THE YEARS AHEAD

New strip mill promises brighter future for thousands *BY MARGARET STEWART*

THERE is no problem of mass unemployment in Britain today as there was before the war. But there are still black spots. Scotland has 72,000 jobless, 3.3% of all its workers. In Greenock and Port Glasgow the proportion is as high as 8.1%. The national average is 1.6%.

New light industry has brought 50,000 jobs to the Clyde valley since the war. But Scotland is still over-dependent on heavy industry.

The steel industry is playing a vital part in redressing the balance, and bringing hope of an industrial renaissance. Colvilles, one of the leading British producers, are going ahead with ambitious plans to raise their steel-making capacity from 2·3 to 3·3 million tons a year. Their most important and imaginative project will produce 500,000 tons of steel sheet and 175,000 tons of light plates a year. It is in two parts – a hot strip mill at Ravenscraig and a cold reduction mill at Gartcosh, 8 miles away.

This is a completely new development for Scottish industry, traditionally a user of heavy steel. The Scots hope that plentiful supplies of sheet steel will attract manufacturers of cars, car accessories, refrigerators, washing machines, cookers, and office furniture. A canning industry could follow. Rootes and the British Motor Corporation have already decided to go north.

Snowball effect

Sir Andrew McCance, chairman of Colvilles, told me: "But for the strip mill, these motor car firms would not have elected to come to Scotland. They will bring a lot of ancillary trades, and the cumulative effect is bound to be substantial."

Mr. John Lang, steel union leader, chairman of the Scottish Board for Industry, and treasurer of the Scottish TUC, said there had been many encouraging enquiries.

The Government's decision to lend Colvilles £50 million for Ravenscraig caused controversy in Parliament. Labour spokesmen said it was wrong to use public money for the benefit of shareholders. Conservative back-bench MPs attacked the loan as "bastard socialism".

But the Scot-in-the-Street was spoken for by Mr. Lang: "We don't care a damn where the money comes from, so long as our men don't have to walk the streets looking for jobs."

Plans are going ahead in spite of the recession in the car industry. What extra employment will result? Perhaps another 1,500-2,000 steel workers will be needed. The car developments should provide upwards of 10,000 jobs directly. The Pressed Steel Company is expanding its plant near Paisley to produce bodies for cars and commercial vehicles. A new development by the accessory firm of Rubery Owen, scheduled for Bathgate, will mean work initially for about 200. And there will be a "snowball" effect from these developments within the next few years.

MARTIN: "A dandy layout"

100% success

It was stimulating to visit the Ravenscraig and Gartcosh sites. The men on the job impart their own infectious enthusiasm.

William Banks, general manager of Ravenscraig, has watched its progress for three years. "I have a great deal of faith in the future of the steel industry, and of our company in particular," he said. "I believe we shall have 100 per cent success."

More steel-making capacity is being installed to supply the new mill. Ravenscraig will be one of the first large plants in this country to adopt the "L.D.-A.C." process, one of the most modern and economic methods of making steel, using oxygen and lime.

About two-thirds of the output of the hot strip mill will be sent to Gartcosh to be cold-rolled into high quality sheet.

There is a strong reason for building at Gartcosh. Nine hundred men work there in the old hand mills of Smith and McLean. Their livelihood depends on steel.

Progress at Gartcosh, as at Ravenscraig, is well up to schedule. George Martin, the "boss", told me that American experts to whom he showed the plans had commented, "That's a dandy layout."

At first 500 men will be employed, rising to 1,000. Most of Smith and McLean's men will be found work. "We will go out of our way to place them in suitable jobs," I was told.

The two men who will manage the hot and cold strip mills are both in their early thirties. James Watson, manager-designate of the hot strip mill at Ravenscraig, is 34. His opposite number at Gartcosh, James Easton, is 31. Both have been through Colvilles' 'sandwich' courses at the Glasgow Royal College of Science and Technology. Both are brimful of confidence in themselves, their industry and the future of Scotland.

WATSON: "Finest thing that ever happened"

Says Watson: "This is the finest thing that has ever happened to me." Says Easton: "The prospects couldn't be better."

The articles on these two pages form part of an advertisement published in the Press by the British Iron and Steel Federation

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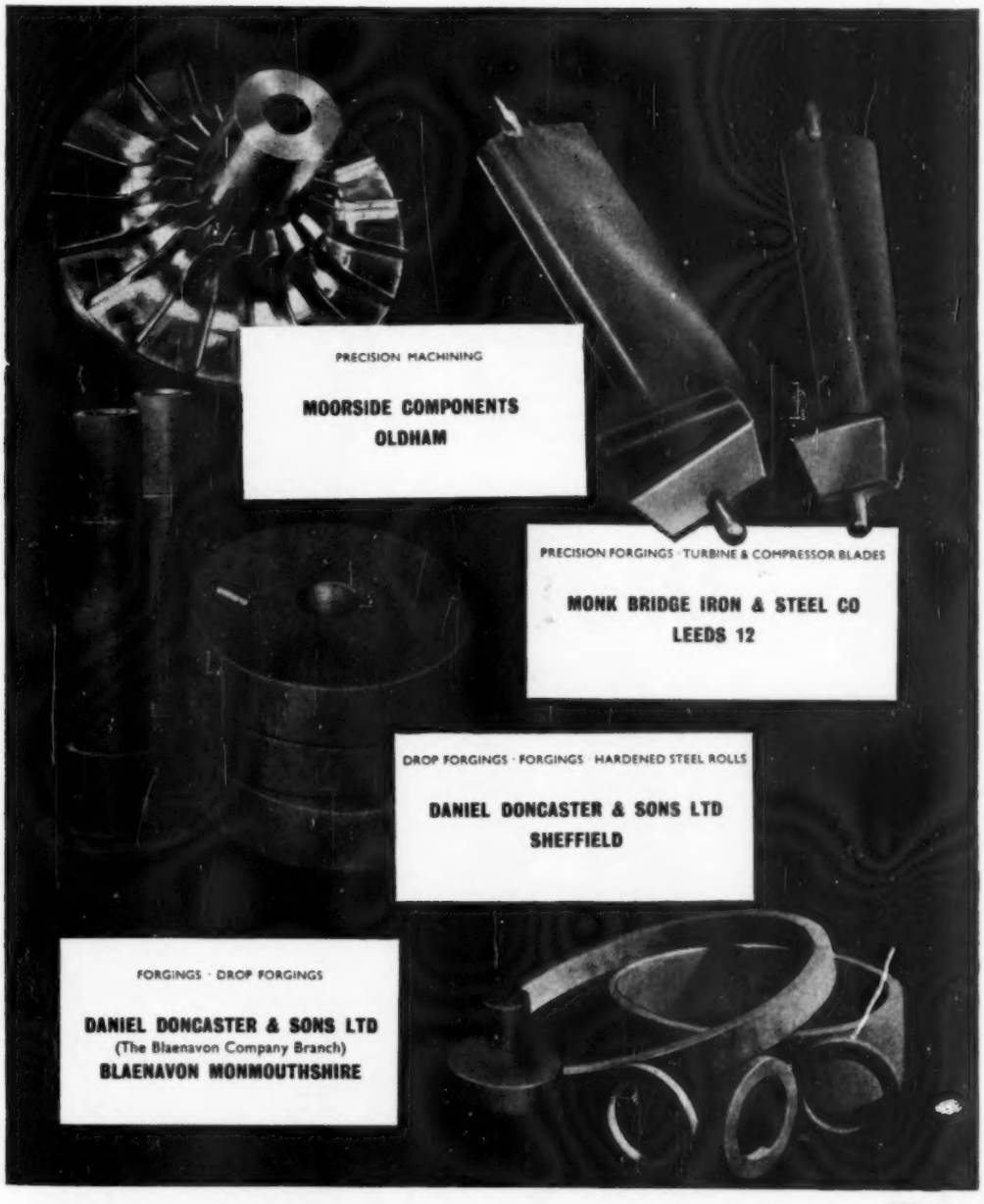

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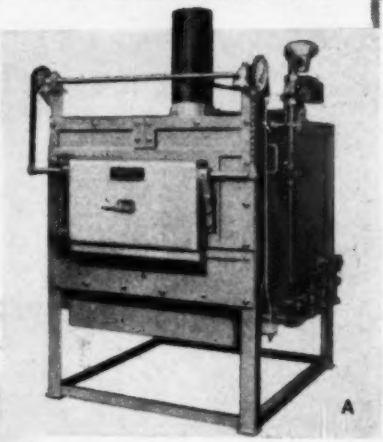
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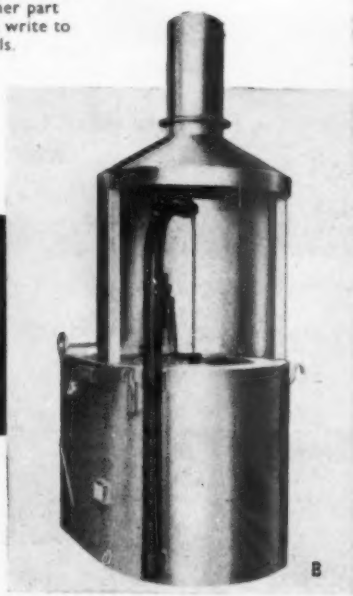
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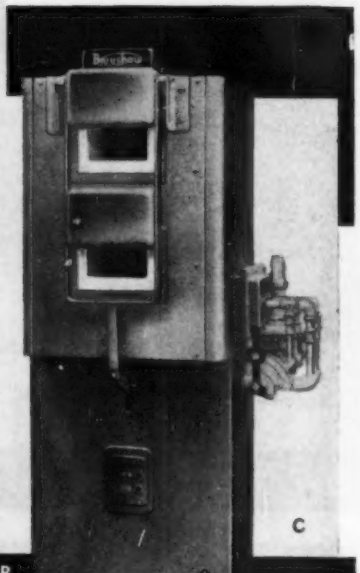
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What the illustrations show:

A.—'Hynor' Oven Furnace.
B.—Liquid Bath Furnace.
C.—Twin Chambered High Speed Steel Furnace.

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2. Finish forge in Maxipresses on the same heat.

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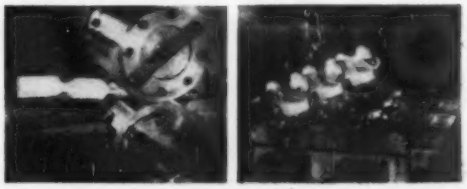

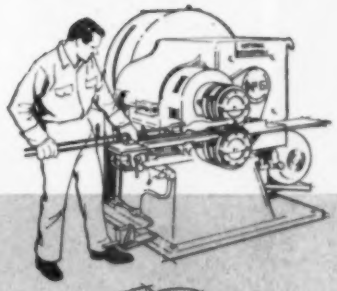

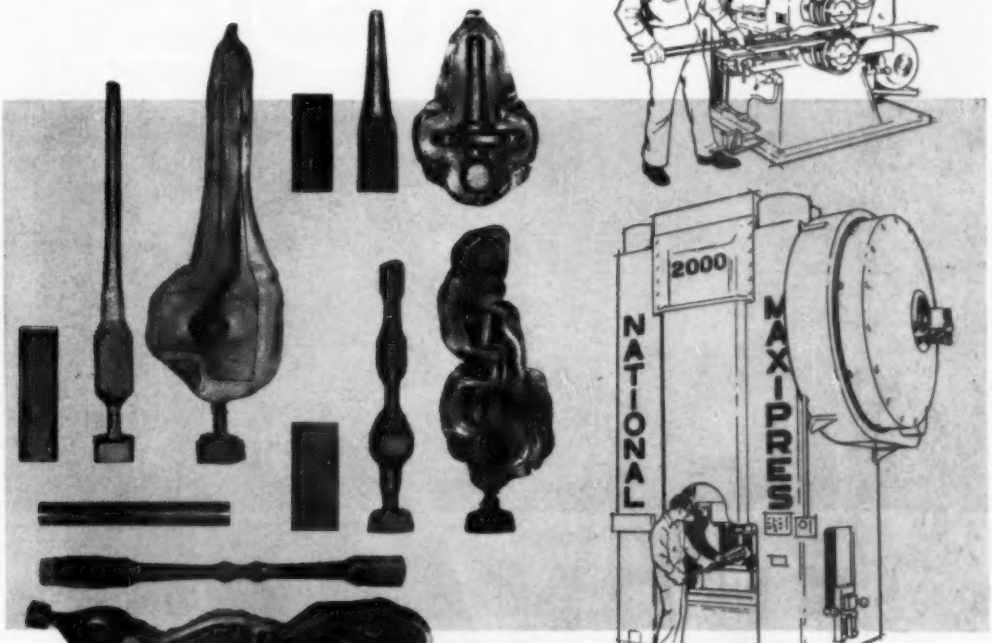

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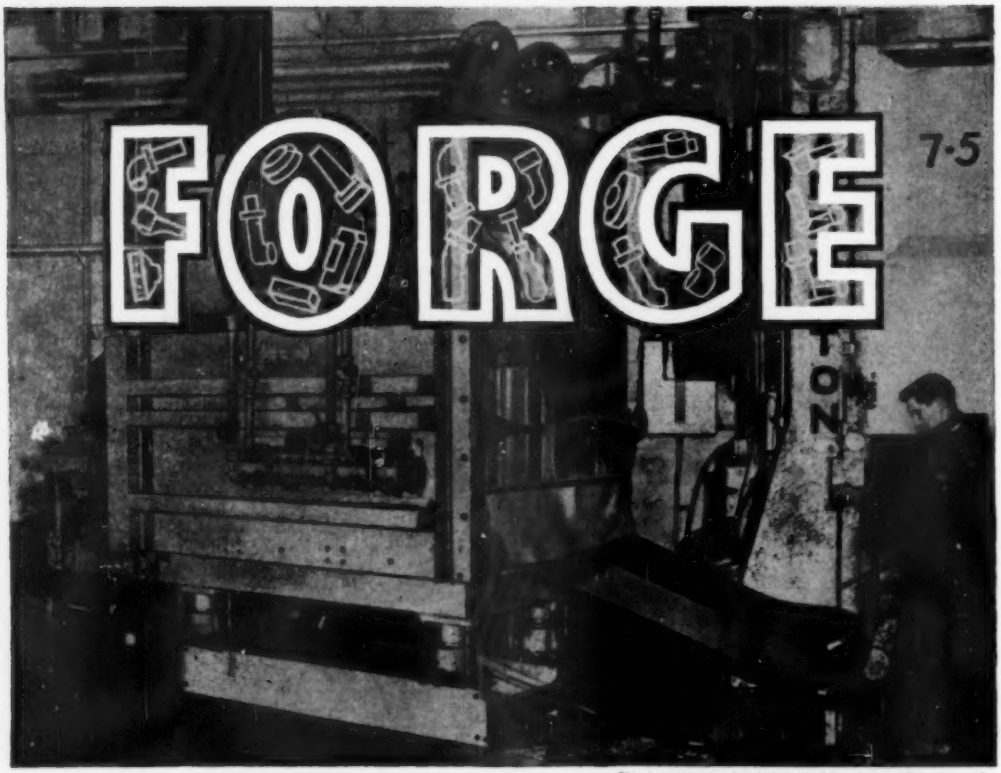

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A wide range of standard or tailor-made mechanised forge furnaces is available, Pusher Type - Rotary Hearth Type - Conveyorised Bar End Heating, with GGC scale-free heating system or fired by gas or oil. The photograph above illustrates a small Thermic Magazine Feed Pusher furnace fired by CC Burners. Output: 7-cwts. of small billets per hour.

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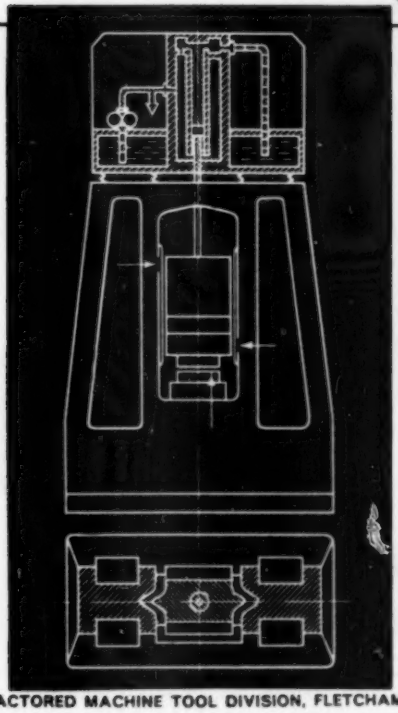

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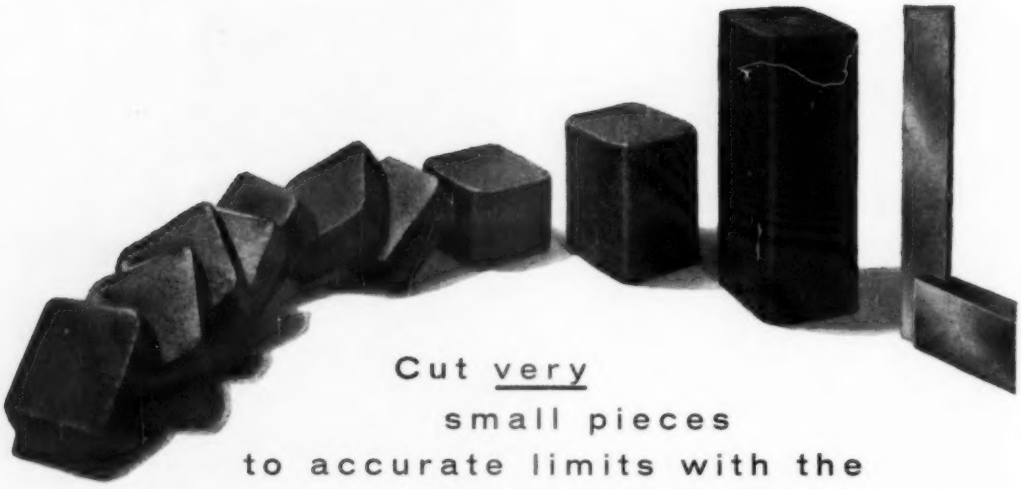

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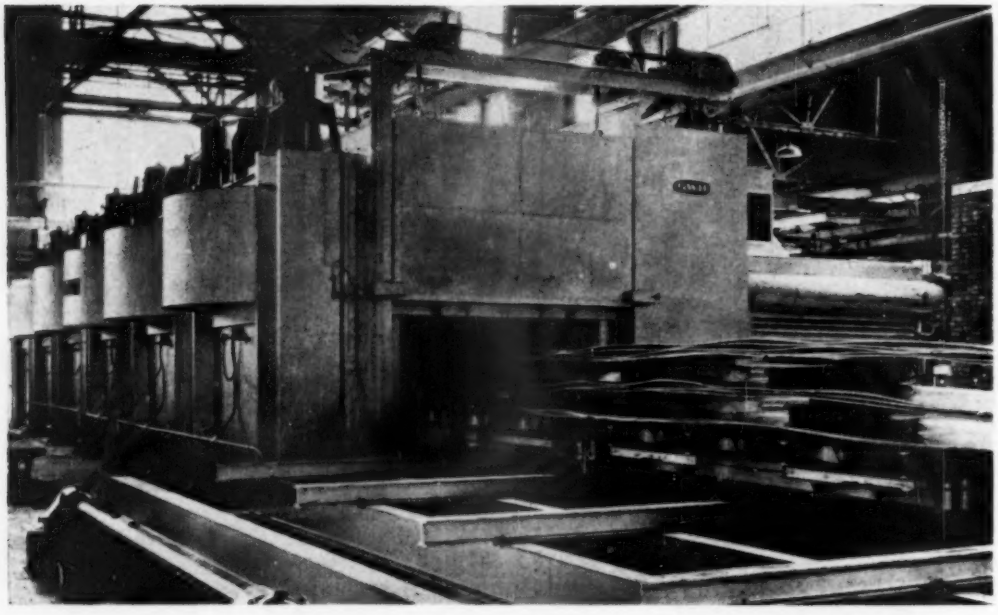

Batch Type Furnace installed, with slow cooling chamber at right, at Northern Aluminium Co. Ltd. for annealing aluminium slabs prior to cold rolling.

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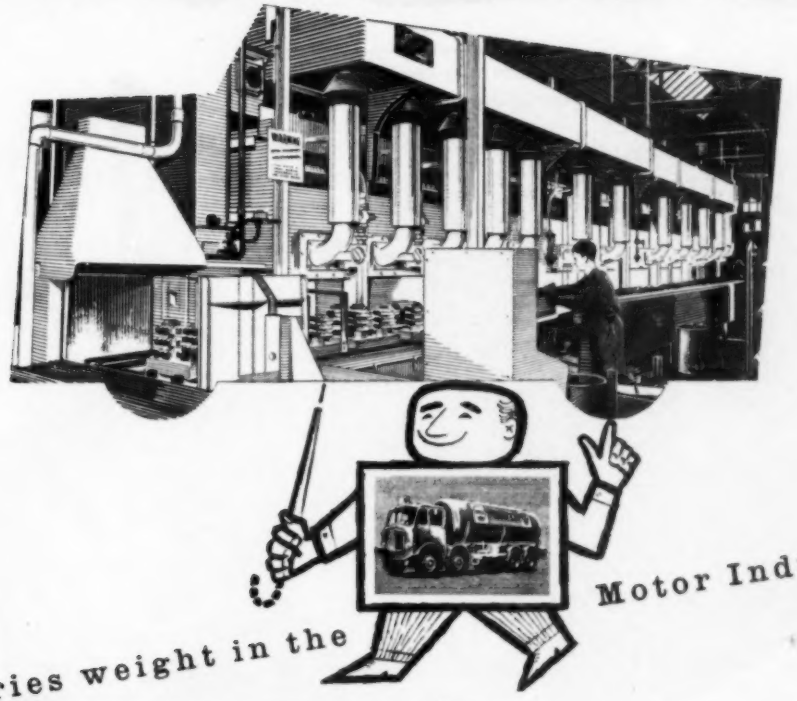

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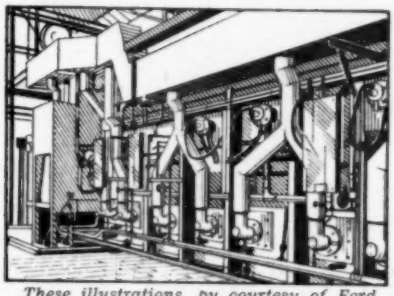

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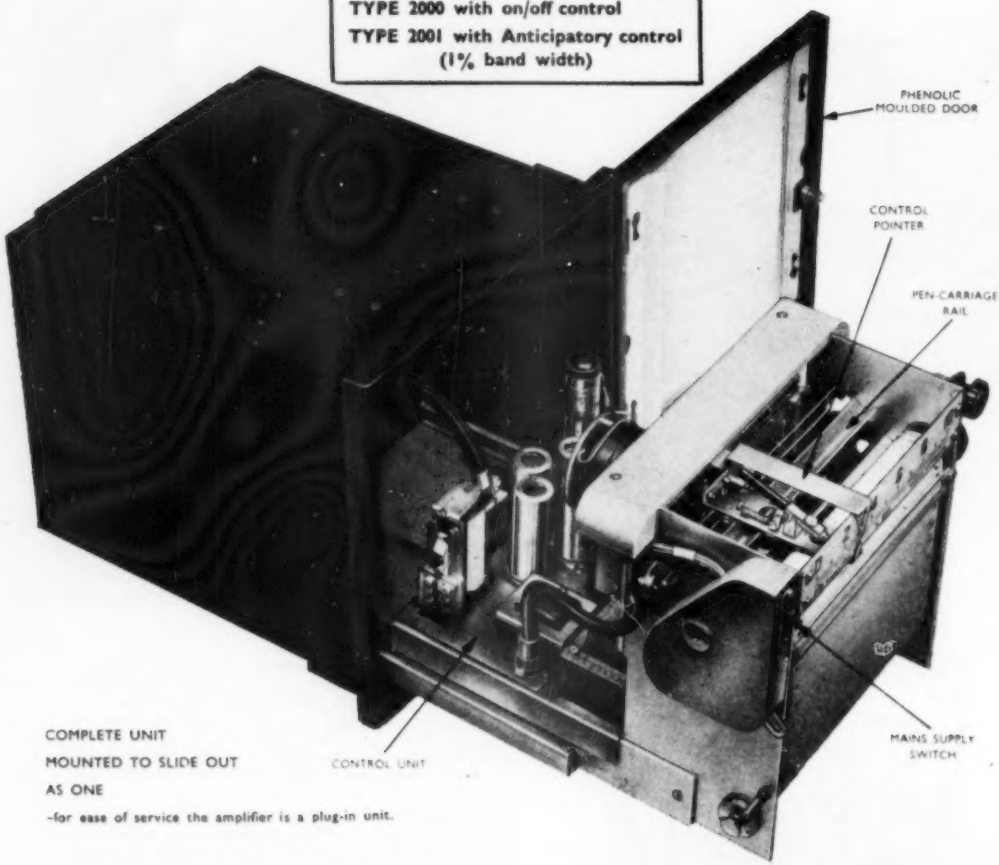

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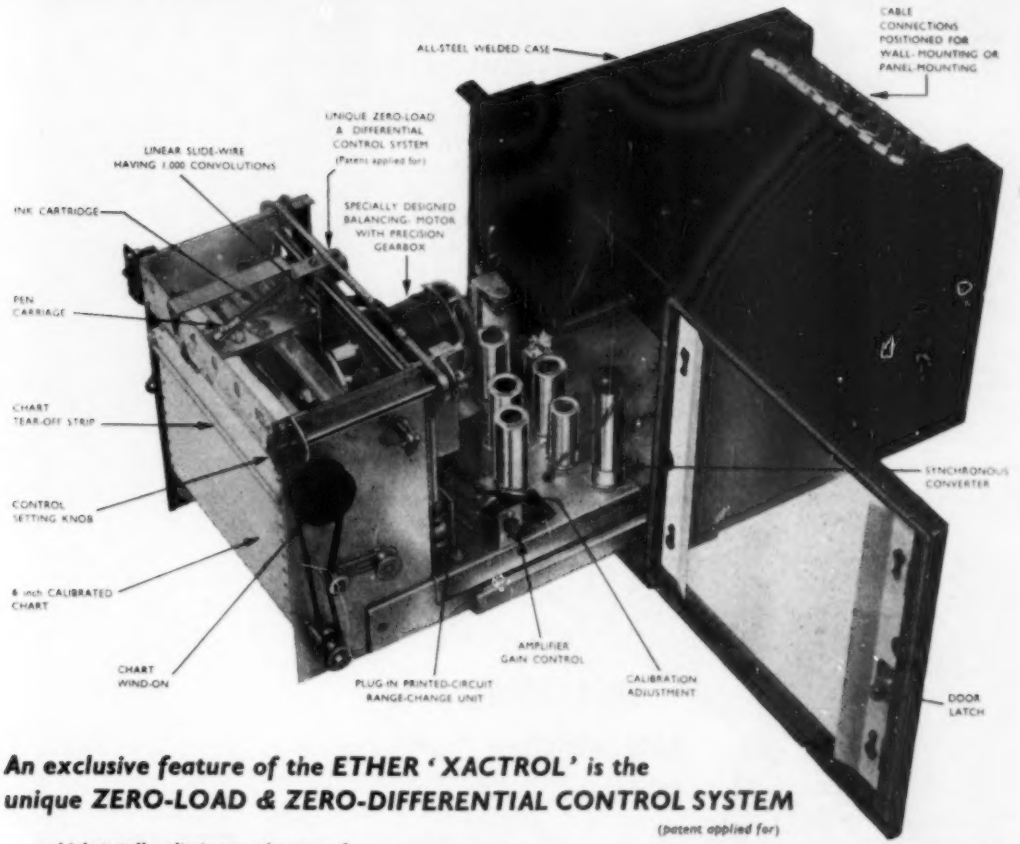

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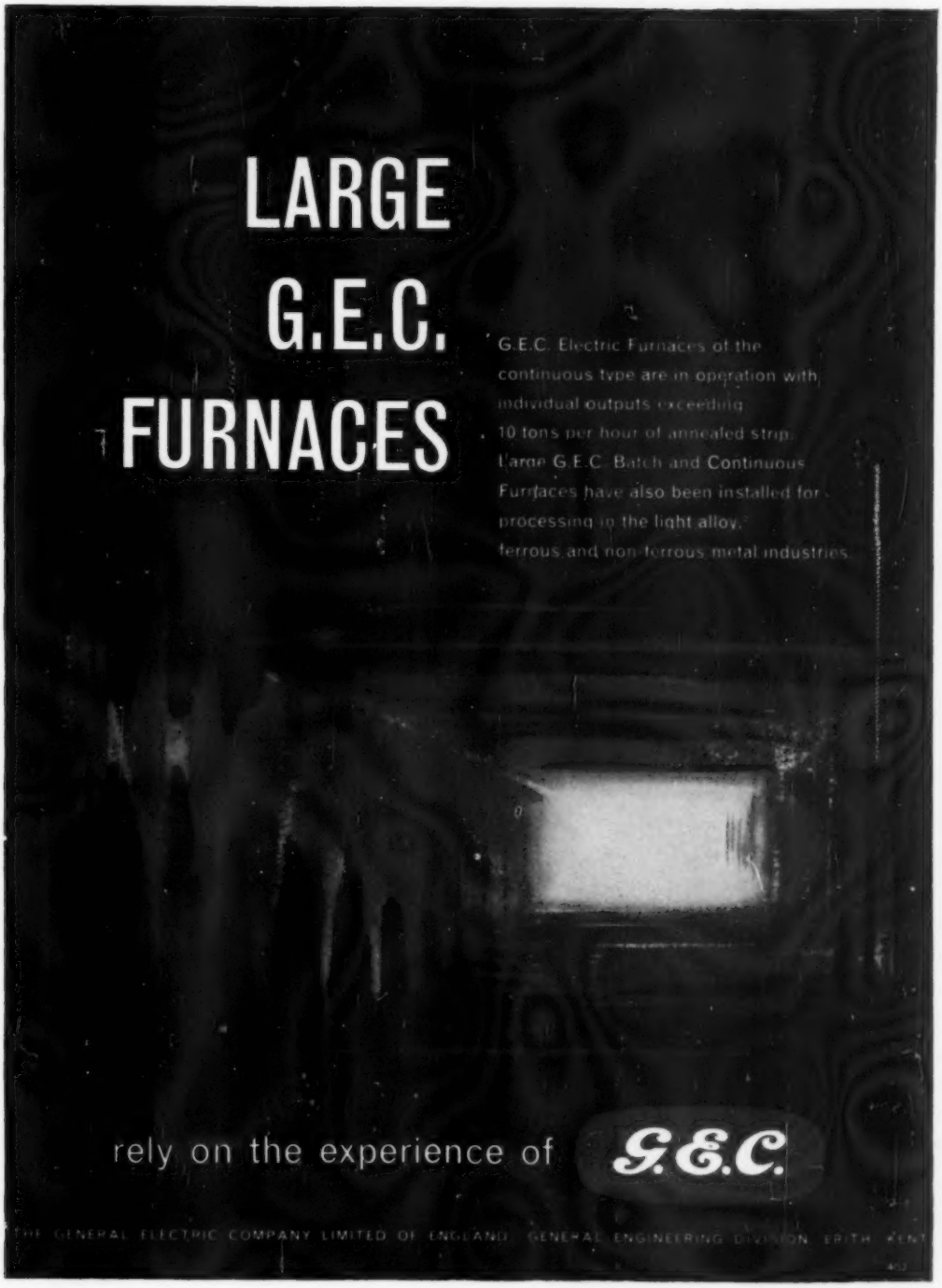
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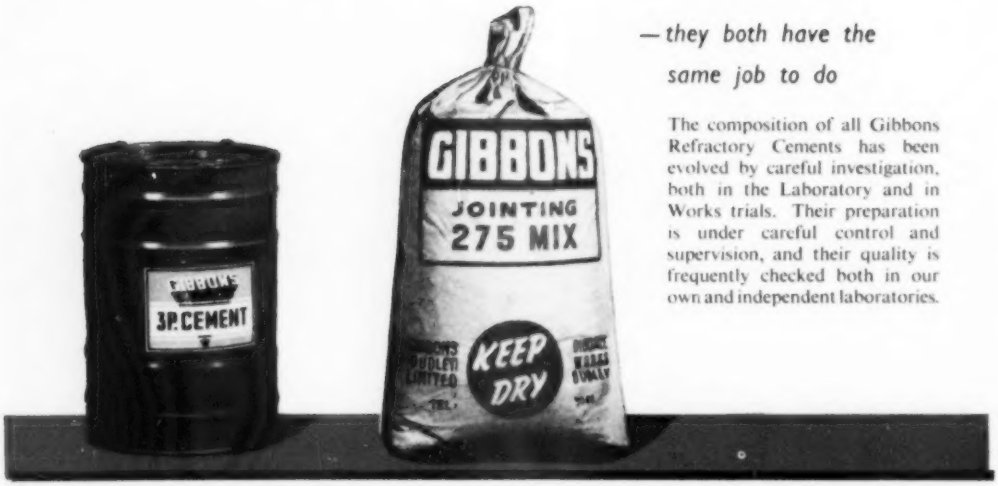

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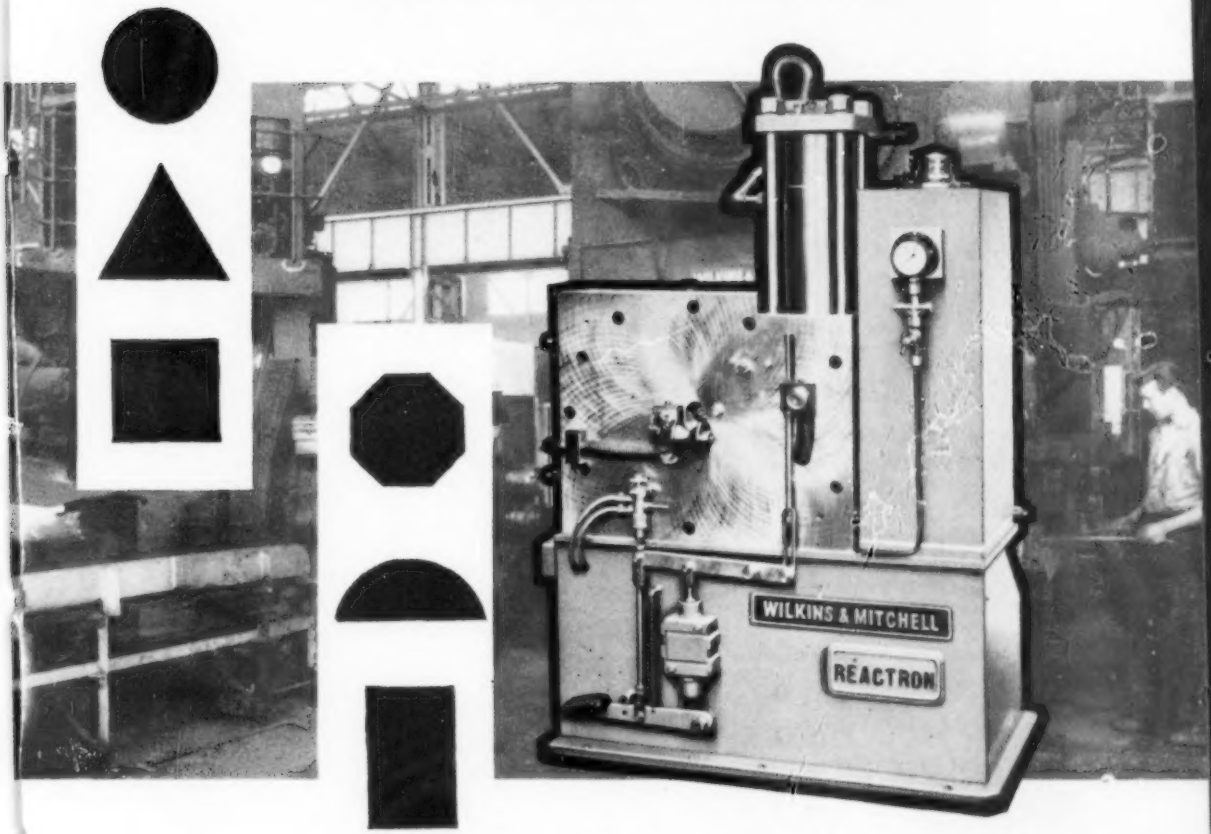

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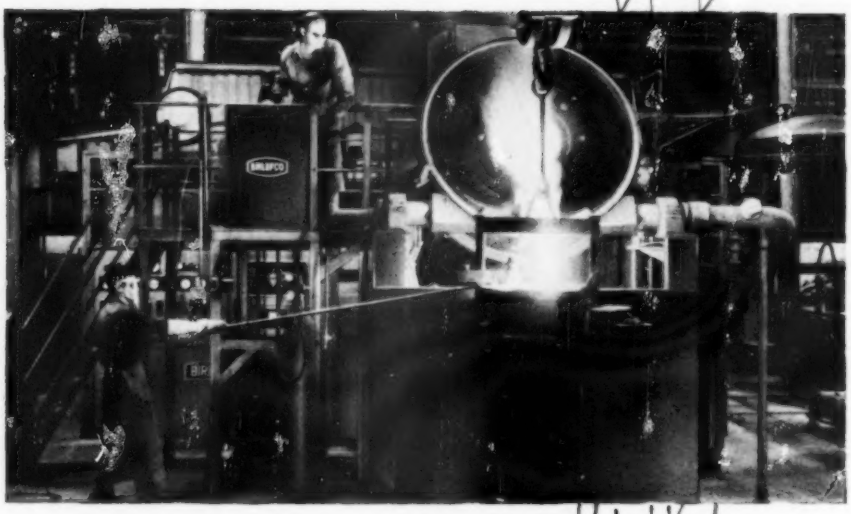

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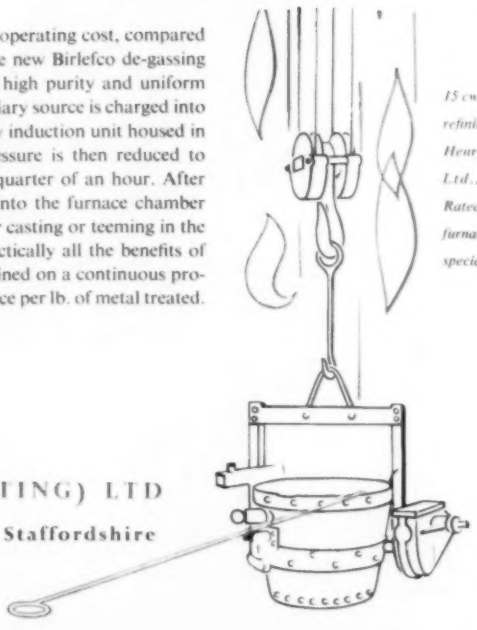

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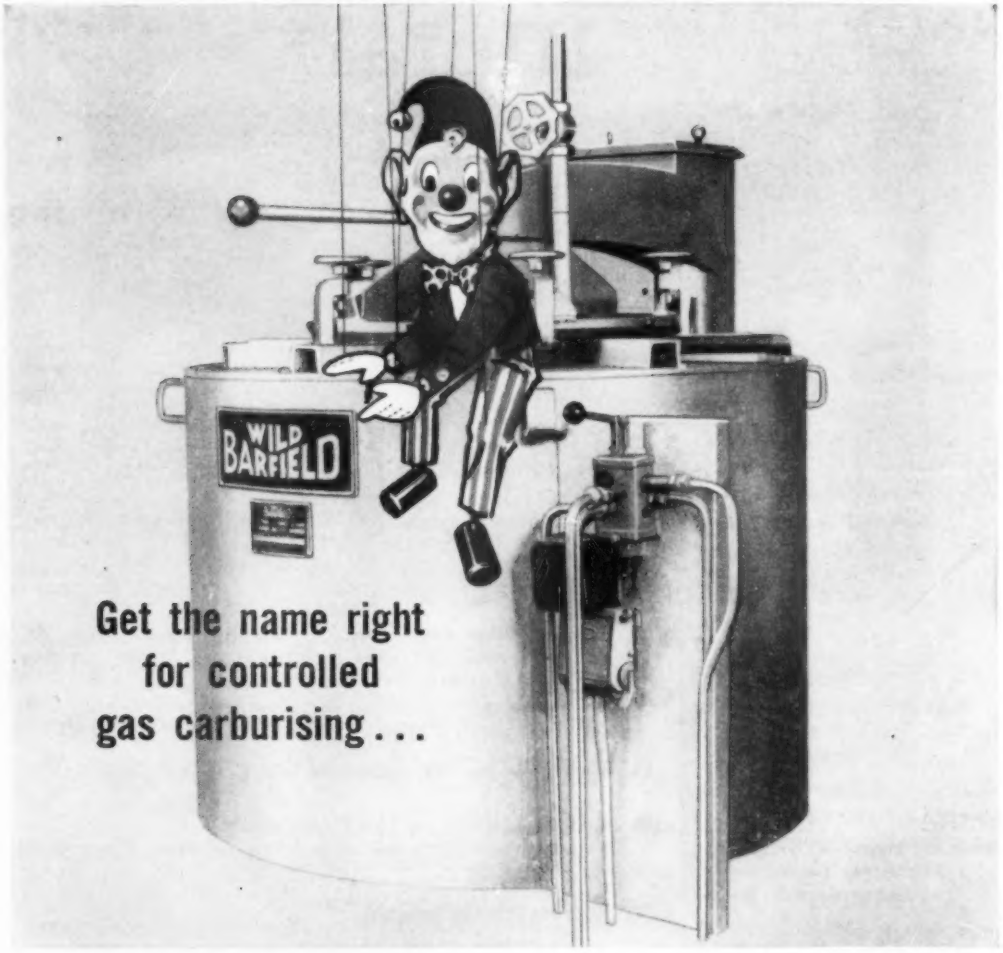

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W.B. 101

Steelmaking 1961

EXACTLY a year ago in this column we noted that the British steel industry had begun, rather belatedly indeed, to recognize the threat of lower cost steel being produced abroad by the improved converter methods. It had also become clear that the higher maintenance and capital costs of the open-hearth furnace were no longer justified on the grounds of unique quality of product since similar qualities were being achieved by the new methods. Nevertheless, the timely application of tonnage oxygen to the open hearth undoubtedly helped British steelmakers to maintain a competitive position in 1960, and has shaped the pattern of steelmaking in this country for at least the next few years.

Looking at the general scene, the outstanding developments completed last year were the new four-high plate mills at Consett and at South Durham's new South Works, backed in the case of South Durham by a new melting shop and slabbing mill. Other developments completed included expansion of coke, sinter, pig iron and rolling capacity at John Lysaght's Scunthorpe Works and the reconstruction of the Hallside Works of Colvilles for the production of special and alloy steels, including a new billet mill.

According to the British Iron and Steel Federation, last year the industry spent about £130 million on new plant and development and raised its production capacity to just under 26 million tons. It is expected to spend at least £150 million during 1961 on development which will increase crude steel capacity to 27 million tons. Work is well forward with the two new strip mills of Richard Thomas and Baldwin and Colvilles, and the first units are expected to be in operation by the end of the year.

Apart from the strip mill developments, major developments due for early completion are the additions of universal beam mill stands to the heavy section mills at Colvilles' Lanarkshire Works, South Durham's Cargo Fleet Works and the United Steel Companies' Appleby-Frodingham Works. Billet capacity will be expanded by the completion of the two new rotor furnaces for pre-refining and the commissioning of the new continuous billet mill at Richard Thomas and Baldwins' Redbourn Works. Other important developments include the new melting shop at Consett with two L.D. and two Kaldo vessels and the completion of South Durham's new South Works by the addition of a blast furnace and further steelmaking capacity.

The capacity of the industry is being increased from just under 26 million ingot tons in 1960 to 27 million ingot tons in 1961. 1962 will see a further sharp rise in capacity to 30 million tons—equivalent to the level of demand anticipated in 1965—due primarily to the great expansion of sheet capacity provided by the completion of the two new wide strip mills. Further developments are likely to raise capacity to over 34 million tons in 1965. It has already been decided to spend at least a further £450 million by 1965.

This then is the shape of things at the moment, and it may be said that it looks rather more hopeful than at this time last year. However, a very alert eye still needs to be kept on foreign developments, and a willingness to explore promising new methods on a production basis is most necessary. The installation of both L.D. and Kaldo at Consett is an excellent example of realism in the study of alternative methods and should provide a most useful basis for comparison. It is to be hoped that more of such ventures will be inaugurated and the results made generally available and applied on a national scale.

The need for greater productivity

G. H. SIMMONS, B.A., F.I.I.Tech.

Last month, in our editorial, 'Drop forging today,' we discussed some of the topics dealt with at the last Technical Convention at Droitwich of the National Association of Drop Forgers and Stampers. Although the following paper does not deal with the technical problems of forging, it is believed that the general interest and importance of the subject today justifies its inclusion in these pages. Mr. Simmons, who presented this paper at the conference, is the divisional director of S. J. Noel-Brown & Co. Ltd., consultants to industry, hospitals and local government

IT IS OBVIOUS that a paper dealing with productivity will be worthless unless we are quite clear what we mean by productivity. Reading in the popular and technical Press about redundancy, automation, operational research and such foreign-sounding words as cybernetics and ergonomics, too often we tend to do one of two things—either we just skim lightly over the articles or we read them closely, without really understanding their meaning, and become depressed and perhaps even angry. Many of our industrial disputes today stem from this dangerous ignorance of new words and slogans. I have heard a foreman, responsible for the organization and control of a large department in a pottery, express satisfaction at the level of productivity in his department having risen by some 10,000 units. I have also heard it said that productivity techniques will not work in a particular industry because it is not highly repetitive.

What, then, is productivity? Productivity is a measure of the results, in terms of production, obtained from the use of our basic resources. It is, in fact, the ratio of output to input and may be expressed in man-hours or machine-hours, in output per unit of space or material. Increased production, on the other hand, may be obtained by increased effort, longer hours of work, more machines and using more material. But increased production obtained by any one, or all, of these methods can only be obtained at a heavy cost, whether in the health of our work people or the rapid deterioration of capital investment, resulting in a falling off in rate of saving and build-up of capital reserves. The object of an industry should be to provide the goods and services required with the minimum consumption of real resources. Too often, and too glibly, the reason given for being in business is to make money, or to make a profit, whereas I feel that this is purely secondary. Money is only an evaluation, a token measure, and it is not as tangible as our more real resources.

Responsibilities of management

It is management's responsibility to provide goods and services through the productivity of an enterprise rather than through wasteful consumption of the effort and skills of manpower, or ineffective use of machines consuming excessive quantities of material in badly designed and badly planned factories. I do not use the word production, because there are many industries which have no product but to whom the aim of higher productivity is equally valid. Such industries are retailing, transport, hotels and catering, hospitals and certain aspects of clerical and administrative work. The measure of their success is related to the service they provide.

I have specified management's responsibility, but we must be clear what management is before we can define the field of responsibility and detail a course of action. Management is the organization and control of human activity directed towards specific ends. Management's first job is to obtain the facts, otherwise the whole function of management is a misnomer and, like a house built on sand, is without a firm and positive foundation. A plan is formulated and the moves on the chessboard of the factory, or work-place floor, must be co-ordinated so that they interlock like a perfectly timed and executed military manoeuvre. Once the pattern has been established it must be controlled and, finally, those taking part should be made to feel that they 'belong' in all this activity; they must be shown their contribution to the end product, whether it is a manufactured article or a service. Management must motivate, not necessarily by means of a financial incentive, but by stressing the worth, by the recognition of individual

skills and by the removal of frustration and 'dead-endedness.'

Packaged deal

We have defined management and outlined its responsibilities, can we now present it with a packaged deal, a 'do-it-yourself' kit, which contains a blueprint, a job specification and a set of tools? Unfortunately, no packaged deal can supply the perfect answer. The aim of the manager of any enterprise should be the basic work content of a product or operation, and this is a perfect condition which never occurs in practice. It may be approached, particularly in process industries such as chemicals and refineries, but usually operation times are far in excess. The basic work content is the time a product would take to manufacture or an operation take to perform, if the design or specification were perfect, the process or method perfectly carried out, and with no loss of working time from any cause whatever (except legitimate rest pauses for the operator).

Shortcomings

Defects in design or specification may make impossible the economical use of plant or machinery and the manpower operating that machinery. There may be a complete lack of standardization, or control of variety, resulting in short batch work, and long and costly periods of machine change-over down time. Unrealistic quality standards may cause re-operations or rectifications owing to unnecessarily close tolerances or over wide limits may call for additional machining to clean up.

There are many examples of inefficient methods adding to the work content—the wrong type or size of machine, the specified or designed rate of production not being worked, the wrong tools, a bad layout which influences throughput and the working methods of the operator which cause wasted movement, time or effort. All the items in the excess work content may be attributed to management weakness. This is true even of bad working methods on the part of the operator where these are due to lack of training and supervision.

Ineffective time due to management arises from several factors: (1) a sales policy which demands an unnecessarily large variety of products resulting in time-wasting machine change-overs and lack of opportunity on the part of operators to acquire the necessary skill and rate of working, (2) a time-wasting development or experimental period, owing to designs not being fully developed or finalized; (3) lack of planning, causing operations and operators to be out of phase and out of balance; (4) raw materials supplies not arriving to schedule; (5) inefficient or negligible maintenance; and (6) poor working environment and a high accident rate.

But what of ineffective time within the control of the man on the shop floor? There is nearly always to be found late starting and early finishing, careless workmanship and the careless behavour of the accident-prone. There is, however, always more ineffective time due to inefficient management than that within the control of workers. By eliminating the causes of ineffectiveness we more nearly approach the basic work content and hence achieve maximum productivity.

The long- and short-term plan for the forging industry

If the basic work content is the ideal and the shortcomings outlined are the hazards to be overcome, what of the tools? How long must we wait for success and how much is it all going to cost? There are two courses open, a short-term plan and a long-term plan, and in discussing these I shall, as far as possible, point my remarks to the drop-forging and stamping industry.

Firstly considering the long-term plan—this will undoubtedly be costly and the benefits may be marginal. Through research and development we can improve the basic processes, and capital expenditure can be made on new plant and equipment. New high-speed forging plant can be purchased, the potential of larger furnaces can be examined together with research into alternative methods of heating, and factory layout revised to enable a more widespread use of mechanical-handling devices. A considerable amount of time and money can go into a relatively straightforward programme on die design, multiple-impression dies with good finish, longer die life and precision tools for producing them. It must also not be forgotten that these improvements will necessitate an extensive programme of labour education—new skills in the tool-room and a better appreciation of new plant and equipment. I think the industry must also face up to a parallel approach of more money for these higher-skill demands.

To simplify and improve the product and reduce the variety is a continuous programme and not to be defined as either long or short term. There are many examples in other manufacturing industries where insufficient care is paid to product design, and slight or major deviations from the original prototype result in excess costs on tool or diemaking, wasteful stores areas and out-of-balance production processes. I am sure this is true of the drop-forging industry in instances where too much reverence is paid to the slogan 'the customer is always right.' I feel it is significant that the scrap value of steel can be as low as 20% of the

purchase price of the bar or billet. In many plants, however, the losses are heavier where the volume of scrap takes up useful factory space.

The short-term plan should prove vitally worth while because the results are achievable now and capital expenditure is negligible. By improving existing methods of plant operation, the planning of work and the use of manpower, a positive approach can be made to achieving maximum machine and labour utilization. The average machine utilization in this country is 42% and it is not a matter for congratulation to run a machine for 44 h. a week which only represents 26% of a basic work content of 168 h. Every blow of the hammer should produce useful work; how close does the British industry approximate to the American rate of 60%, where the hammer rate has been reduced to 10 or 12 blows? Manpower efficiency should be raised by proper planning and servicing—by a reduction of walking time. If an operator walks 8 yd. from furnace to hammer, to press and return to furnace for a production rate of 120 pieces/h., this is equivalent to 24 miles per week or, assuming three miles per hour, an 8-h. day spent achieving nothing except a considerable increase in product cost.

Manpower effectiveness can be increased in other directions by good housekeeping, elimination of time looking for tools and a positive safety programme—wearing of goggles, gloves, safety boots and helmets. Such factors as improving the ventilation, reducing the heat and the cumulative effect of weight will assist labour recruitment and reduce labour turnover. Finally, the direct financial incentive based on sound assessment by work measurement must not be overlooked.

Work study

Immediate and far-reaching results can be achieved for the short-term plan by the use of work study, by employing the twin techniques of method study and work measurement. *Method study* is the systematic recording and critical examination of existing and proposed ways of doing work, as a means of developing and applying easier and more effective methods and reducing costs. *Work measurement* is the application of techniques designed to establish the time for a qualified worker to carry out a specified job at a defined level of performance. It is to be noted how closely interlocked are these twin techniques, since work measurement relies on a job specification which can only practically be provided by method study.

The sequence of moves in the application of method study are logical and therefore cannot be transposed. Firstly, the problem is defined under the heading of technical, economic or human and the facts relating to the problem are obtained by an objective recording and examination. There are several possible means for recording, from a form of shorthand using symbols to the frame-by-frame accuracy of the cine camera. Three-dimensional scale models may also be used, the whole purpose of the particular method employed being to provide a positive and dynamic presentation instead of vague and often exaggerated hearsay.

The job specification

A better way of doing the work is developed after the most searching and critical examination and this is applied after joint discussion and mutual agreement with everyone, however remotely concerned. Work study cannot be applied in a vacuum and we cannot ignore the part human beings play, nor their strengths and weaknesses, their doubts and fears.

Method study provides the blueprint of the job specification and work measurement provides the dimensions. How much can we save, how long is this order going to take, when can the customer expect delivery and how much can we give to the operator in the form of a bonus?

The job specification provides the facts on which the efficient enterprise relies—there can be no plan, there can be no production control and there can be no quality standard or human satisfaction without a job specification. Good management relationships cannot exist, much less thrive, where there is chaos, uncertainty, ignorance and frustration. Higher productivity is only possible through the use of management techniques and these in turn are only possible where good management relations exist.

The most important and the most fundamental prerequisite of any practical approach towards achieving higher productivity is the right attitude of mind. We have to overcome the apathy and mental inertia engendered by resistance to change by those two well-worn clichés—'my job is different' and 'we have always done it that way.'

Before the formation of the British Productivity Council in 1954, the Anglo-American Council for Productivity was responsible for the visits of representatives, from both sides of industry, to their counterparts in the U.S.A. and Great Britain, to enable exchanges to be made of industrial know-how. One of these was by the drop-forging industry and the subsequent report contained much that was valuable. The report followed almost the same lines as the present paper, presenting a long-term and a short-term plan for the industry. Now—nearly seven years have passed—how many of the recommendations have been implemented?

Intergranular corrosion of austenitic steels containing 24%Cr and 19%Ni

VLADIMIR CIHAL, PAVEL GROBNER, DR. JAROSLAV JEZEK and DR. RUDOLF POSPISIL

The following electron-microscope study was first reported in Czechoslovakian in 'Hutnické Listy,' 1960, (7). The authors are respectively at the State Research Institute for the Protection of Metals, the Modřansk Engineering Works, the State Research Institute for Materials and Technology, and the United Steelworks, Kladno. The work is dedicated to Dr. Josef Teindl of the Mining High School in Ostrava on his 60th birthday

INTERGRANULAR CORROSION of austenitic, heat-resisting steels, especially chromium-nickel steels, on account of its importance has already been the subject of numerous research works in world scientific literature. Very soon after the discovery of intergranular corrosion, it was established that it is caused by the segregates of carbides (of chromium) on the grain boundaries, and methods of counteracting it were found. Of these, two permanently proved themselves technically effective: firstly, the addition of carbide-forming elements, titanium or niobium for instance, which form carbides at high temperatures which are more stable than chromium carbides; and, secondly, a reduction in the carbon content initially to under 0·08%, and later to under 0·03%.

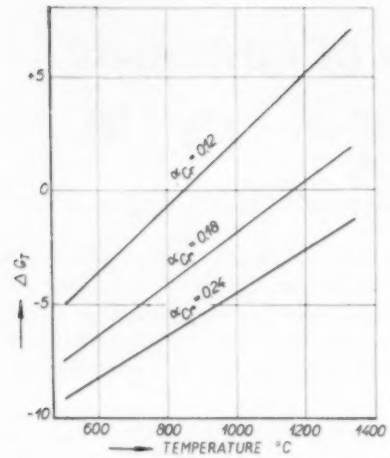
Why the carbide segregates on the grain boundaries should give rise to intergranular corrosion is, in general, still something which absorbs the attention of the metallographer. According to the most widely held opinion, experimentally confirmed on 0·1C-18Cr-9Ni steels, as a result of the precipitation of the chromium carbides the surface layer of the grains is impoverished in chromium down to under 12%, whereby its passivating capacity is so diminished that corrosion occurs. For details of the precipitation of the carbides see our previous reports.[1, 2]

From experience with severely cold-worked stainless steels it is known that considerable internal stress worsens their corrosion resistance. On this basis a second opinion was expressed[3] that the cause of intergranular corrosion lies in the stresses set up by the carbides precipitated on the grain boundaries. As evidence for this opinion it is alleged, firstly, that even steels with an addition of Ti or Nb can be subject to intergranular corrosion if carbide segregates are produced on the grain boundaries as a result of excessively irregular heat treatment, without indeed any proof that these are carbides of titanium or niobium and, secondly, that even steels containing more than 20% Cr can be subject to intergranular corrosion. At the same time it was assumed[4, 5, 6] that in this instance it would be quite impossible in accordance with the first opinion for the chromium content to fall to such an extent as under 12% in consequence of the precipitation of the carbides. The aim of the present work was to investigate the justification for this last-mentioned argument. At the same time we hope to succeed in explaining the problem of intergranular corrosion in 0·1C-24Cr-19Ni austenitic steel.

Thermochemical stability of chromium carbide

The higher chromium content in 0·1C-24Cr-19Ni steel not only restricts the drop in the chromium content below the passivation limit of Cr-Ni steel during the precipitation of carbides on the grain boundaries, but from the theoretical aspect it could also increase the resistance of the $Cr_{23}C_6$ carbides to dissolution in the austenite, and thereby relatively decrease the quantity of carbon in the solid solution at low austenitization temperatures.

This is confirmed by a thermochemical analysis of the stability of chromium carbide, normally depicted in a diagram of the relationship between the free enthalpy of the carbide for 1 gram-atom of carbon (*i.e.* the thermochemical potential) and temperature. The course of the equilibrium plots for the system chromium in the solid solution-carbide of chromium expressed by the equation[7]



1 *Relationship between the thermo-dynamic potential of the carbide $Cr_{23}C_6$ and temperature at activities of chromium in the steel of* 1·12, 0·18 *and* 0·24

$$23/6\ Cr + C = 1/6\ Cr_{23}C_6 \ldots\ldots$$
$$\ldots\ldots\ \Delta G_T = -16{,}380 - 1{\cdot}54 \cdot T \qquad (1)$$

is depicted for three activities of chromium 0·12; 0·18 and 0·24 in the diagram in fig. 1. At the first glance we see that the gradient of the equilibrium plots falls with the increasing activity of chromium, and the plots are displaced to negative values of ΔG_T.

With a view to the practical application of the diagram presented, it is premised that the alloying elements, which form solid solutions with iron to an unlimited extent or on a very large scale, behave almost ideally. Their activity is then roughly equal to their atomic concentration. The possibility of accurate comparison of the activities and concentrations, however, is complicated by the unstoichiometric condition of the chromium carbide, $Cr_{23}C_6$, in which indeed substitution of atoms of chromium by atoms of iron can take place to the extent of up to 30%.[8] Nevertheless, the plot of the thermochemical potential in relation to temperature for chromium carbide shows that with the increasing content of chromium in corrosion-resistant steel there is an increase in the temperature at which this carbide dissolves in austenite. The conclusion derived from the thermochemical analysis led to experiments to test the effect of an increasing austenitization temperature for 0·1C-24Cr-19Ni steel on intergranular corrosion after subsequent, single-stage, precipitation heat treatment.

Chi-Mei Hsiao and H. J. Dulis[9] recently came to a similar conclusion; they maintain that the lowest temperature for the dissolution of the carbide segregates in corrosion-resistant austenitic steels rises with the increasing content of Cr and C in the steel, and falls with the increasing content of Mn and Ni. The effect of Mn and Ni is only slight.

Outside the precipitation temperatures, which cause sensitivity of the steel to intergranular corrosion, the chromium carbides are stable, and their precipitation, which is possible only in the presence of carbon dissolved in the solid solution, will be governed by the rate of diffusion of carbon and chromium on to, and in the direction of, the grain boundaries.

Kinetics of the precipitation of the carbides

The formation of nuclei for the precipitation of carbides is assumed to be so rapid that the kinetics of their further growth is governed only by the diffusion rate. As opposed to the diffusion rate of chromium that of carbon is much higher, so that on the grain boundaries there is a sufficiency of carbon for the formation of the carbides. Above all we shall be concerned with the diffusion rate of C and Cr in austenite, and therefore with their diffusion coefficients. For the subsequent calculations we have based ourselves on the diffusion coefficients compiled in Table 1.

TABLE 1 *Diffusion coefficients of carbon and chromium in austenite, cm.²/sec.*

Temperature, °C.	Carbon	Chromium
550	$1{\cdot}4 \times 10^{-8}$	$8{\cdot}0 \times 10^{-14}$
650	$7{\cdot}7 \times 10^{-8}$	$6{\cdot}2 \times 10^{-13}$
750	$3{\cdot}0 \times 10^{-7}$	$2{\cdot}6 \times 10^{-12}$
850	$9{\cdot}1 \times 10^{-7}$	$8{\cdot}2 \times 10^{-12}$

In view of the linear relationship between log D and $1/T$, the diffusion coefficients[10, 11] for the temperature range of austenite in carbon steel can be used for calculation and application at lower temperatures of 550–850°C. At these temperatures 0·1C-24Cr-19Ni steel is completely austenitic and, according to Wells and Mehl, 20% Ni in austenite causes only a small increase in the diffusion coefficient of carbon. Chromium in an amount of 24% will probably not have a great influence on the diffusion of carbon in austenite, since it forms an ideal solid solution with iron and, in common with iron, has corresponding dimensions of the structural lattice.

For calculations of the diffusion velocities we use the Second Fick Diffusion Law. The stable carbide in 0·1C-24Cr-19Ni austenitic steels is the carbide $(Cr, Fe)_{23}C_6$. Just as in our earlier work[2] with 0·1C-18Cr-8Ni corrosion-resistant steel, we assume that the grain is cubic in shape with an edge length of the cube of 10^{-3} cm., and that the carbide segregates will precipitate on one wall of

the cube.[12] The precipitated chromium carbide lowers the content of C and Cr in the immediate vicinity, on the grain boundary, therefore, to the minimum quantity, or practically to zero. These alloying elements must, during the equalization of the composition between the centre and the boundary of the grain, diffuse through a distance of one wall of the cube of 10^{-3} cm. If we are considering 0·1C-24Cr-19Ni steel, at the concentration of 0·1% C in this crystal, which has a specific gravity of 7·92 g./cm.³, there will be $7{\cdot}92 \times 10^{-12}$ g. carbon. For the formation of $Cr_{23}C_6$ this quantity of carbon requires an optimum quantity of $1{\cdot}27 \times 10^{-10}$ g. chromium. The concentration of chromium in the steel is 24%, so that the grain contains a total of $1{\cdot}87 \times 10^{-9}$ g. C.

On the basis of the assumptions presented we calculate firstly the time required for the diffusion of the required quantity of chromium on to the grain boundary. For this purpose we used the relationship given by L. Jenicek[13] for a quantity of an element S which in time t diffuses through a diameter q at a constant concentration of the diffusing element c_o:

$$S = c_o \cdot q \cdot \sqrt{\frac{D \cdot t}{\pi}} \qquad (2)$$

In our instance $S = 1{\cdot}27 \times 10^{-10}$ g. Cr, $c_o = 1{\cdot}87$ g./cm.³ for 24% Cr, $q = 10^{-6}$ cm.² and D is the diffusion constant in accordance with Table 1. If we insert the values indicated into equation (2), we obtain the values of the times for the diffusion of the required quantity of chromium on to the grain boundaries, which are given in Table 2 and compared with those for 0·1C-18Cr-8Ni steel.

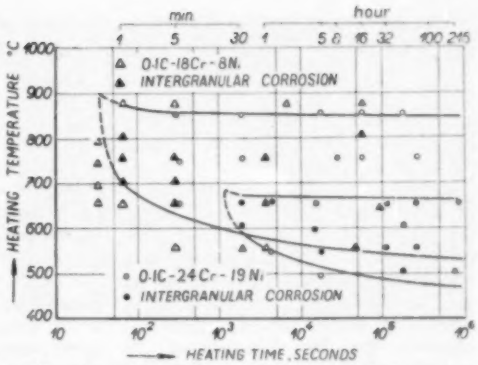
TABLE 2 *Time required for the diffusion of chromium on to the grain boundary for the formation of $Cr_{23}C_6$ carbides at 0·1% C in 24% Cr + 19% Ni steel and 18% Cr + 8% Ni steel*

Temperature, °C.	Time in seconds for	
	0·1C-24Cr-19Ni steel	0·1C-18Cr-8Ni steel
550	$1{\cdot}82 \times 10^5$	$2{\cdot}4 \times 10^5$
650	$2{\cdot}34 \times 10^4$	$3{\cdot}1 \times 10^4$
750	$5{\cdot}56 \times 10^3$	$7{\cdot}35 \times 10^3$
850	$1{\cdot}76 \times 10^3$	$2{\cdot}34 \times 10^3$

With reference to the opinion that intergranular corrosion is caused mainly by the loss of passivation capacity of the zone impoverished in chromium, we shall be concerned with the width a of this zone. This width we calculated from the equation derived earlier:[12]

$$a = 2\frac{S_{Cr}}{c_{oCr} \cdot q} = 1{\cdot}36 \cdot 10^{-4} \text{ cm.} \qquad (3)$$

On completion of the diffusion assumed above,



2 *Diagrammatic representation of the critical zones of tendency towards intergranular corrosion for corrosion-resistant ○ and △ indicate that no tendency to intergranular corrosion was revealed*

i.e. after $2{\cdot}34 \times 10^{-4}$ sec., for instance, at 650°C., on the grain boundary theoretically the whole of the carbon contained in the austenitic grain will have been precipitated in the form of chromium carbides, around which the content of chromium in the solid solution will be reduced, and only at a distance of $1{\cdot}36 \times 10^{-4}$ cm. from the boundary will the Cr content again be 24%. This is the state in which intergranular corrosion can occur in this corrosion-resistant, Cr-Ni, austenitic steel.

During further heating, of course, diffusion will continue in the sense of an equalization of the concentration of chromium. For the calculation of the concentration of chromium, firstly at the lower limit of the passivation of iron, we shall consider 12% Cr to be the minimum content in the iron capable of effecting its passivation.

The time required for raising the concentration of chromium on the grain boundary from zero to 12% cannot be determined entirely accurately, but an approximate calculation is possible on the basis of certain simplifications. For the calculations we use the computation of the Second Fick Law:

$$\frac{c}{c_o} = \frac{1}{2}\left[1 - G\left(\frac{x}{2\sqrt{D \cdot t}}\right)\right] \qquad (4)$$

Where c is the concentration of the diffusing element at a distance x from the plane of constant concentration c_o, which in our instance is $c = 0{\cdot}95$ g./cm.³ for 12% Cr;

c_o for a content of 24% Cr in the grain = 1·87 g./cm.³ and for 18% Cr c_o = 1·425 g./cm.³;

G is the Gauss function;

D is the diffusion constant; and

t is time.

The assessment was carried out on the basis of the Seith table,[14] which gives directly values of

$$\frac{x}{2 \cdot \sqrt{D \cdot t}}$$

for various ratios of c/c_0. In our instance the ratio varies from 0 to 0·51 for 0·1C-24Cr-19Ni steel (from 24% to 12%) and from 0 to 0·66 for 0·1C-18Cr-8Ni steel (from 18% to 12%), and likewise the distance x is not constant. For approximate calculation we used average ratios of c/c_0 of 0·26 and 0·33. For the distance x we write the length of the edge of the cube, 10^{-3}, of the cubic grain mentioned earlier for the calculation of the time required for the diffusion of the appropriate quantity of chromium on to the grain boundary. For the average ratios of the concentrations of 0·26 and 0·33 the values of

$$\frac{x}{2 \cdot \sqrt{D \cdot t}}$$

are 0·796 and 0·675, so that we can calculate the approximate time t in which the concentration on the grain boundary is increased to 12% Cr (Table 3).

Within this last-mentioned period the concentration in the centre, and on the boundary, of the grain is equalized, but within the whole grain the concentration will be lower by the proportion required for the precipitation of the carbides, so that it will fall to 22·4% Cr in the 0·1C-24Cr-19Ni steel and to 16·6% Cr in the 0·1C-18Cr-8Ni steel.

TABLE 3 *Time required for raising the Cr concentration on the grain boundary to 12% at 550–850° C.*

Temperature, °C.	Time required for			
	0·1C-24Cr-19Ni steel		0·1C-18Cr-8Ni steel	
	Seconds	Hours	Seconds	Hours
550	$5·05 \times 10^6$	1,404	$6·9 \times 10^6$	1,915
650	$6·46 \times 10^5$	222	$8·9 \times 10^5$	248
750	$1·54 \times 10^5$	43	$2·1 \times 10^5$	59
850	$4·86 \times 10^4$	14	$6·7 \times 10^4$	19

The time required for this equalization of the concentration we calculate only approximately, if we use the average ratio, $c/c_0 = 0·5$, since in reality this varies from 0 to 1. For the ratio of the concentrations of 0·5 the value of

$$\frac{x}{2 \cdot \sqrt{D \cdot t}}$$

equals 0·477, and the approximate calculated time t, within which equalization of the concentration of chromium takes place at various temperatures, is shown in Table 4.

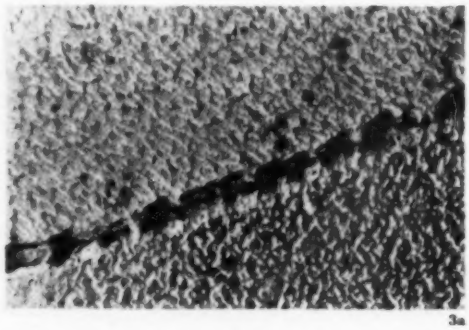
All these calculations are only approximate and subject to certain assessment, but even despite these disadvantages it is possible to suggest, in view of the small difference in the calculated numbers of hours between 0·1C-24Cr-19Ni and 0·1C-18Cr-8Ni, that the higher content of chromium in the first steel does not prevent the tendency of this steel to intergranular corrosion.

TABLE 4 *Approximate time required for equalization of the concentration of chromium in the austenitic grain at various temperatures*

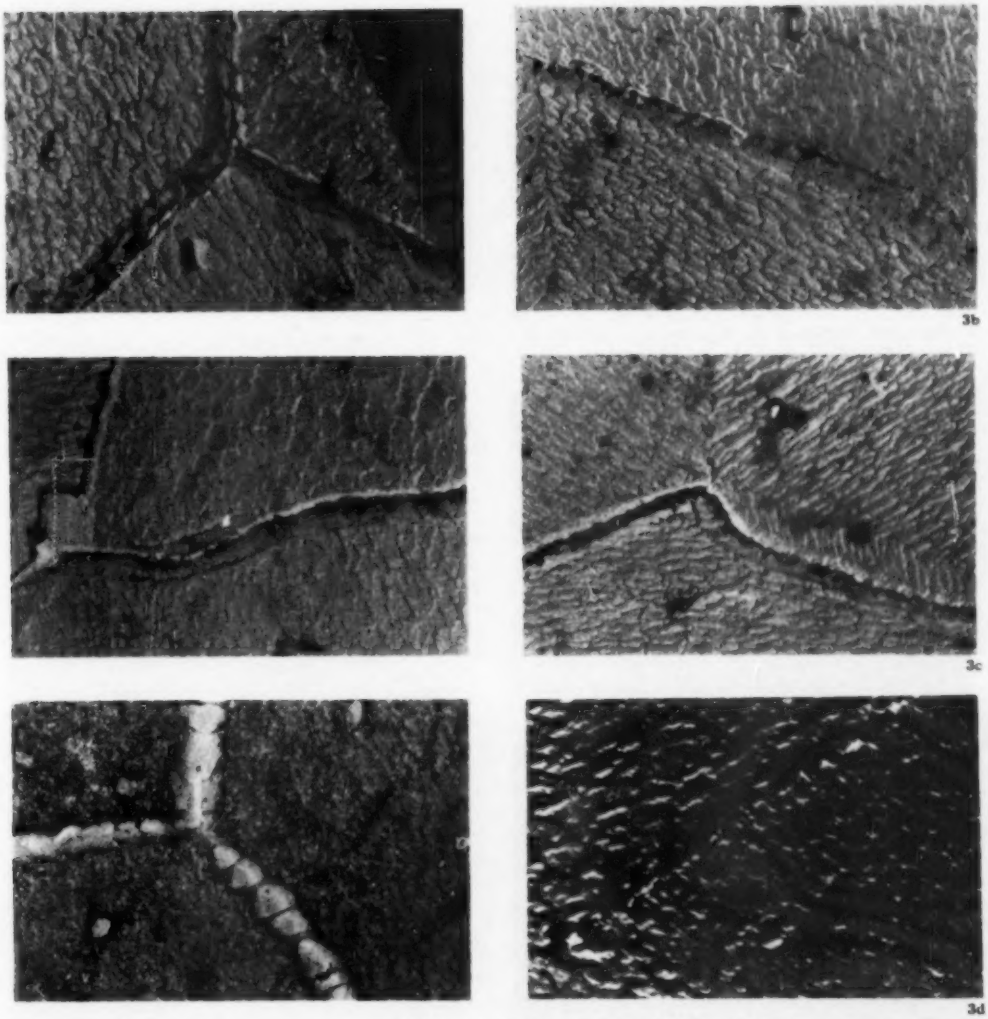
Temperature, C.	Time of equalization of the concentration in hours
550	3,800
650	495
750	108
850	37

Experimental results

For the experiments we chose the austenitic, heat-resisting 0·1C-24Cr-19Ni steel ČSN 17255, Poldi AKC, which is sometimes also used as a corrosion-resistant steel cast from a high-frequency, basic furnace; its composition was 0·09% C, 0·04% Mn, 1·5% Si, 23·2% Cr and 18·7% Ni. After milling, from the small ingots strips were forged of dimensions 25 × 6 mm. On the specimens of this steel we investigated the tendency to



3a



3b

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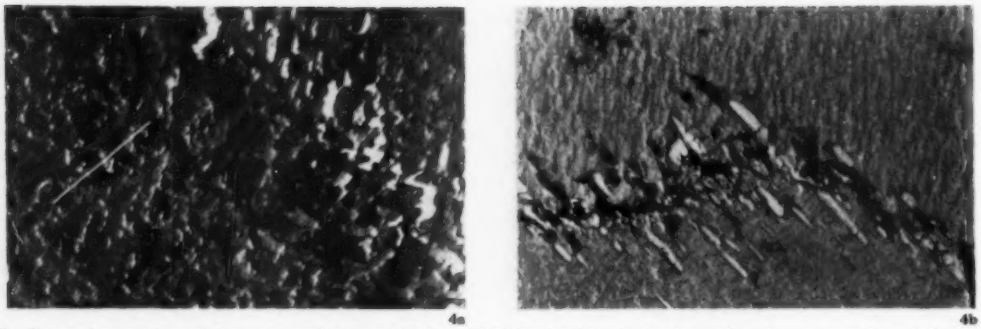
3d

3 *Electron micrographs of extraction replicas of* 0·1C-24Cr-19Ni *steel austenitized at* 1,100°C. *for* 30 *min. and quenched in water and precipitation heat treated at* 550°C. *for:* **a** 5 *h.;* **b** 32 *h.;* **c** 72 *h.;* **d** 810 *h., with cooling in air* × 7,000

intergranular corrosion after various heat treatments, and the distribution of the chromium carbides in the structure of the steel after precipitation heat treatment. The various heat treatments of the individual specimens are shown in Tables 5 and 6, together with the results of the tests for intergranular corrosion. Firstly the specimens were austenitized at 1,100°C., and then they were precipitation heat treated within a range of temperatures from 500–850°C. In order to be able also to compare the influence of the austenitization temperature, in addition some further specimens were heat treated at temperatures from 950 to 1,250°C. with temperature increments of 50°C., and subsequently precipitation heat treated.

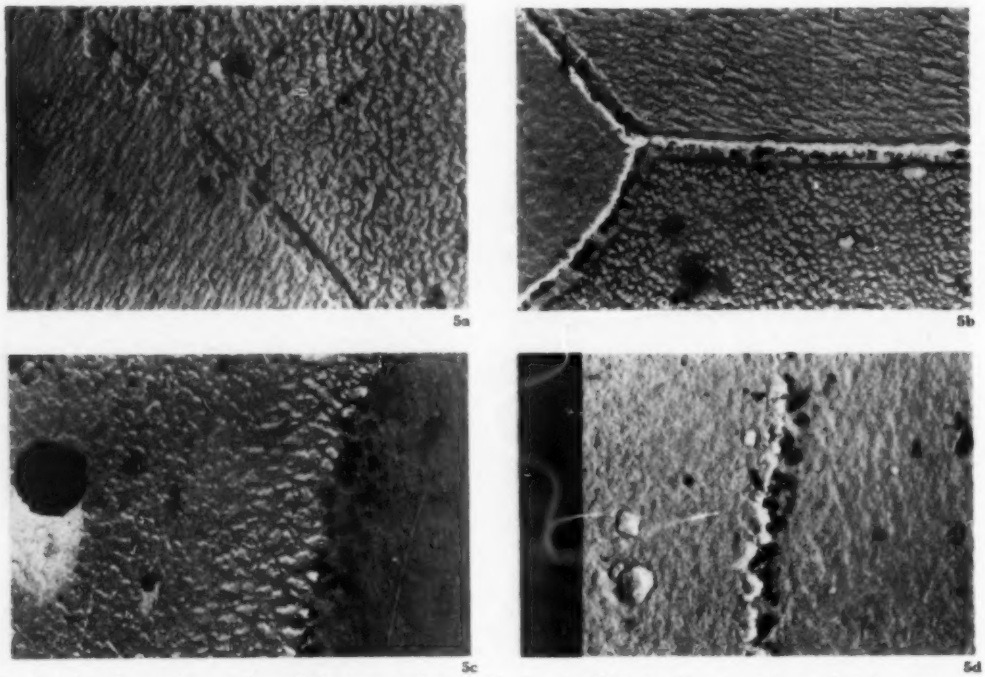
Sensitivity to intergranular corrosion

The intergranular corrosion test was carried out by means of 48 hours' boiling of the specimens, which had previously been ground with 000



4a 4b

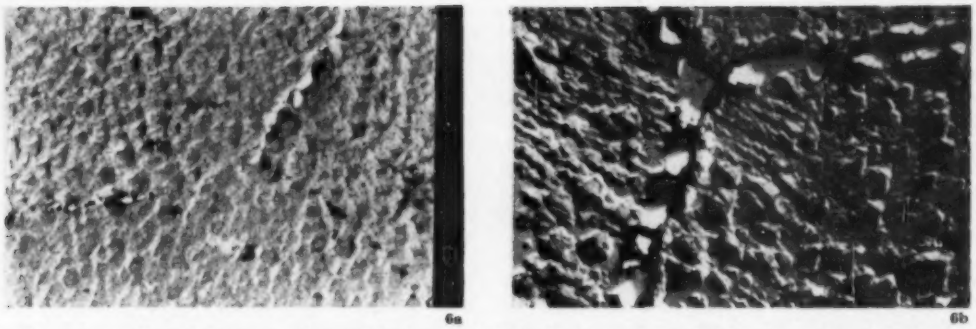
4 *Extraction replicas of the same steel, austenitized at 1,100°C. for 30 min. and quenched in water, and precipitation heat treated at 650°C. for 72 h. and cooled in air* × 7,000



5a 5b 5c 5d

5 *A series of extraction replicas of the same steel austenitized at 1,100°C. for 30 min. and quenched in water, and precipitation heat treated at 750°C. for times of:* **a** *6 min.;* **b** *65 min.;* **c** *8 h.;* **d** *72 h.* × 7,000

abrasive paper, in an aqueous solution of sulphuric acid and copper sulphate (100 g. conc. H_2SO_4 + 100 g. $CuSO_4 \cdot 5H_2O$, made up to 1,000 ml. with water) in the presence of copper swarf, in flasks with repeated cooling. Into each flask were inserted four specimens interspersed with glass wool and a handful of copper swarf, and immersed with solution, so that about 100 ml. of the solution were poured into each flask. The presence of metallic copper in the solution of sulphuric acid and copper sulphate maintains the oxidizing reduction potential of the solution at a constant value of + 0·4 V. from the start to the end of boiling. With specimens less inclined to intergranular corrosion, therefore, the period of a drop in the oxidizing reduction potential of a solution without



6 *Extraction replicas of the same steel austenitized at 1,100°C. for 30 min. and quenched in water and precipitation heat treated at 850°C. and cooled in air:* **a** *for 30 min;* **b** *for 48 h.* × 7,000

metallic copper does not occur; in such a solution, as a result of the reduction of Cu^{++} to Cu^{+} through the reaction on the grain boundaries, there is a drop from +0·75 V. to + 0·4 V. Metallic copper also makes possible the reduction of Fe^{+++} to Fe^{++}, and the intergranular corrosion also takes place at the maximum possible rate. The intergranular corrosion test-pieces were prepared in duplicate, and assessment of the intergranular corrosion was carried out by bending the specimens, and the occurrence of intergranular cracking was judged visually (Tables 5 and 6).

In agreement with the thermochemical hypotheses, according to the results obtained, the degree of intergranular corrosion increases with the rise in the austenitization temperature, for a continuously greater quantity of carbon passes into the solid solution, which during subsequent precipitation heat treatment is liberated in the form of an unbroken network of chromium carbides on the grain boundaries. The most developed intergranular corrosion takes place after precipitation heat treatment at 500–600°C. After a longer period of heating at 650°C. and at higher temperatures, intergranular corrosion no longer occurs, despite the fact that the presence of the chromium carbide $Cr_{23}C_6$ on the grain boundaries was established by means of X-rays and electron microscopy.

TABLE 5 *Effect of precipitation heat treatment on the intergranular corrosion of 0·1C-24Cr-19Ni steel austenitized at 1,100°C. for 30 min. and quenched in water**

Time of heat treatment	Heat treatment temperature, °C.					
	500	550	600	650	750	850
5 min.				–	–	–
30–65 min.		–	(+)	(+)	–	–
4–5 h.	–	+	+	+		–
8–16 h.					–	–
32 h.		++		(+)		
48–50 h.	++		++			–
72 h.		+++		(+)	–	
216–267 h.	+++			(+)		
320–358 h.			(+)	–		
810–860 h.	++	++				
1,920 h.		+				

*The degree of intergranular corrosion rises from + to +++; – denotes complete absence of intergranular corrosion, and + very slight intergranular corrosion.

TABLE 6 *Effect of the austenitization temperature on intergranular corrosion after subsequent precipitation heat treatment at 600°C. for 48 h. with cooling in air*

Austenitization (with quenching in water) for 30 min. at °C.	Precipitation heat treatment	Result of the intergranular corrosion test*
950		–
1,050		–
1,100	48 h. at	++
1,150	600°C. with	++
1,210	cooling in air	+++
1,250		+++

*Assessed by the same method as in Table 5.

At these precipitation heat-treatment temperatures the equalization of the chromium content on the grain boundaries and within the grains is much more rapid in the steel with the higher chromium content. The difference in the sensitivity to intergranular corrosion between 0·1C-18Cr-8Ni steel and 0·1C-24Cr-19Ni steel is represented graphically in the diagram in fig. 2, and is shown to be substantially greater than would correspond to the theoretical calculations presented earlier. The higher chromium content, similarly to a higher titanium content, displaces the most favourable precipitation heat-treatment temperature for the occurrence of intergranular corrosion towards lower temperatures.

Distribution and morphology of carbides in 0·1C-24Cr-19Ni steel

In the type of steel studied, just as in 0·1C-18Cr-8Ni steel, precipitation occurs of only

one of the three known carbides of chromium, namely $Cr_{23}C_6$. For the corrosion and mechanical properties of these steels the decisive factors are not only the morphology and dimensions of the precipitating carbides but also above all their location in the structure. For this reason, for the investigation of the manner of the precipitation of the chromium carbides in specimens after various heat treatments, we used simplified extraction replicas, which we studied further in the electron microscope.

Even after a short period of heating the 0·1C-24Cr-19Ni austenitized steel at a temperature of 550°C., on the grain boundaries fine particles of chromium carbide are apparent, which with the extension of the period of heat treatment grow both in size and quantity. This is evident from electron micrographs of specimens mechanically polished and etched in a solution containing 5 g. $FeCl_3$, 50 g. HCl and 100 g. H_2O. In the extraction replicas, which were lightly shadowed with chromium, apart from the impressions of the etched structure of the 0·1C-24Cr-19Ni austenitic steel, are included segregates of the carbides precipitated on to the grain boundaries of two grains and at the junction of three grains (fig. 3).

With the rising temperature of the precipitation heat treatment a rapid growth in the carbides takes place, as is shown by fig. 4, obtained from a specimen heat treated at a temperature of 650°C., or in figs. 5*c* and 5*d* for specimens heated at 750°C. In certain instances the chromium carbides attain such dimensions that they can no longer be differentiated by an extraction replica, and leave on the replica only normal impressions (figs. 3*d*, 4 and 6*b*). Morphologically the chromium carbides in 0·1C-24Cr-19Ni do not differ from the carbides which precipitate in 0·1C-18Cr-8Ni steels. The crystallographically developed carbides form oriented chains or strings. It is interesting that, just as in 0·1C-18Cr-8Ni steel, the segregates of the precipitated carbides are located on the grain boundaries, always predominantly in one of the adjacent grains. The diffusion velocity of chromium in 0·1C-24Cr-19Ni steel at a temperature of 650°C. and above is already so high that the initial impoverishment in chromium arising in the vicinity of the precipitated carbides is very rapidly evened out. In consequence the chromium carbides have morphologically developed shapes.

Application of chromium-nickel steel ČSN 17255

Steel ČSN 17255 is a basic type for austenitic heat-resisting steels, the Cr and Ni content of which has been changed somewhat as a result of development. As a general rule this 0·2C-24Cr-19Ni steel is employed as a heat-resisting steel at temperatures up to 1,100°C. for components of annealing, tempering and enamelling furnaces, in glassworks, for the construction of recuperators, etc. This steel cannot operate continuously at 600–900°C., since embrittlement occurs through the precipitation of the σ-phase, even though this is relatively slow. At these temperatures it can be used, for instance, for annealing and case-hardening vessels and similar objects, which can readily be annealed from time to time, so that the precipitation of the σ-phase never attains a destructive degree.

Austenitic steel of this type has outstanding anti-corrosive properties, which may be employed with advantage in pickling baths for metals and alloys, for instance for pickling drums, baskets, hooks, tongs and other implements. It helps the corrosion resistance of this steel if its carbon content is kept below 0·1%. In this study, therefore, 0·1C-24Cr-19Ni steel was chosen; with this carbon content it is also used as a corrosion-resistant steel.

Conclusions

From the results of this work it follows that, even in this 0·1C-24Cr-19Ni steel, as a result of the precipitation of chromium carbides on to the grain boundaries, in the surface layer of the grains the chromium content can fall below the passivation limit, thereby creating the necessary conditions for the intergranular corrosion of this steel. The theory of internal stress can therefore be rejected both here and also for 0·1C-18Cr-9Ni steel with or without Nb; for this steel we have already shown earlier[1, 2] that the artificially precipitated segregates on the grain boundaries are not titanium carbides or niobium carbides but the chromium carbides, $Cr_{23}C_6$.

Recently the drop in the content of chromium in austenite during the precipitation of chromium carbides in corrosion-resistant steels was established from direct X-ray spectral analysis[15] by means of a device for micro-analysis with an electronic probe. This equipment[16] makes it possible at the present time to carry out accurate quantitative analysis and local identification of the lattice structure roughly within a volume of a cubic micron, as has already been described earlier.[17] By this direct, local analysis unambiguous proof and confirmation was provided of the impoverishment in chromium of the grain boundaries of corrosion-resistant steels in the vicinity of the precipitated carbides.

Acknowledgments

The study was carried out in the G. V. Akimov State Research Institute for the Protection of Metals in Prague with the co-operation of the

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Technique for determination of diffraction patterns

JOHN A. BENNETT, of the Metallurgy Division, U.S. National Bureau of Standards, has developed a 'photomicrometer' for measuring X-ray diffraction patterns.

The primary difficulty encountered in the X-ray measurement of stress or strain in metals is the lack of sensitivity due to the small change in diameter of the diffraction ring with change in stress and the diffuse nature of the ring, which makes precise measurement of its diameter difficult. The Knorr-Albers microphotometer, although accurate, suffered from the disadvantages that the chart record was jagged, making it difficult to locate the maximum, and the time required to obtain measurements was excessive.

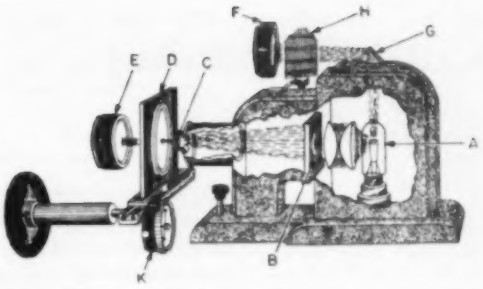
In order to overcome these disadvantages without decreasing the accuracy obtained on the microphotometer, the new instrument was designed for measuring the patterns directly.

Details of the apparatus

The apparatus is essentially a densitometer with a dial indicator to measure movement of the film. A novel feature is a curved slit in the optical system so arranged that the light beam which scans the diffraction ring is an arc having the same radius of curvature as the ring. In this way the entire ring over an arc length of 15° is automatically integrated to determine the point of maximum density.

After passing through the film, the light is received on a barrier-layer-type photocell. This is connected in a compensating circuit with a galvanometer and a second cell illuminated by direct light from the source. Thus, fluctuation in the galvanometer reading due to changes in intensity of the source are eliminated. A fairly sharp maximum in the galvanometer reading is observed when the diffraction ring on a typical pattern is scanned, so that the results can be reproduced within 0·01 to 0·02 mm.

In operation the film carriage is moved until the maximum density of the diffraction ring is located, and the reading of the dial indicator recorded. The carriage is then shifted until the light beam is positioned on the inside edge of the calibrating ring; the indicator reading is again noted. The difference between these readings and the known radius of the calibrating ring gives the radius of the diffraction ring.



Schematic diagram showing operation

Light from source (A) is focused by condenser on to a curved slit (B). The image of slit reduced 5 to 1 by 24 mm. objective. (C) is focused on to film of holder (D) and light through film is measured by photocell (E). Variations in source are compensated by (F) through prism (G) and screen (H). Movement of film holder is measured by (K).

A comparison of measurements from the Bureau's instrument with those obtained from conventional microphotometer records showed both results to be nearly identical in actual magnitude of the average pattern radius as well as in precision (as indicated by the scatter of individual readings). Use of the new instrument over several months period has saved about 25% of the total time or 15 to 20 minutes per pattern.

Intergranular corrosion

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United Steelworks in Kladno and the State Research Institute for Materials and Technology in Prague; the authors wish to thank all these organizations for their support in carrying it out.

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Cold deforming

H. J. MERCHANT, F.I.M.

The following article is based on a talk given by the author to the Australian Drop Forging Association at the third annual convention last year. The author is on the administrative staff of Australian National Industries Ltd. and was formerly works manager of its forging subsidiary, Australian Forge and Engineering Pty. Ltd. In recent months, many new developments in cold forming have been the subject of interesting discussions, and Mr. Merchant attended the conference on cold extrusion of steel, held in Sheffield last November by the Institute of Sheet Metal Engineering

COLD DEFORMING is the name chosen to indicate various working operations on metal blanks which have been rolled into bar or flat product forms. Such operations differ in nature but have in common that they are usually performed without the application of heat. Cold deforming of metal involves plastic movement or manipulation, and the basic amount of energy required to effect this varies with different metals and alloys. For example, the softer non-ferrous metals and alloys are easily cold deformed using steel tools, and such metals have been successfully cold deformed for many years.

On the other hand, ferrous metals and alloys, notably the many grades of steel commercially employed, have a much reduced plasticity at room temperatures and are much more difficult to work cold with the tooling materials available to date. This is particularly the case where production involving a drastic change of shape or reduction of cross-sectional area occurs, *e.g.* extrusion or forging, and therefore we find that cold deforming of steel is a comparatively recent development. It was employed during World War II for the production of steel cartridge cases and was the subject of considerable development in Germany in an effort to conserve raw materials. Its intensive development in the U.S.A. dated back about 10 years and in the U.K. about five years. Cold deforming includes the following processes—cold heading, extrusion, impact extrusion, drawing, coining and embossing and cold forging.

Cold heading is a process of cold deforming involving the enlargement of diameter or cross-section and a consequent gathering of metal usually, but not always, performed on the end of a bar or section.

In most cases the operation is performed in a horizontal forging machine involving one, two or even more progressive stages. Usually the metal is fed continuously into the machine in coil form and typical examples are nails, screws, bolts, nuts, special fastenings, balls, tie rod ends, etc.

Extrusion is a process whereby portion of a metal slug or billet contained in a die is displaced through a designed orifice in such a manner as to form a reduced cross-sectional area or a hollow shape. The tooling usually consists of a female-die member and a male-punch member. There are several distinct possibilities as to what shape may be produced. For instance, in ' backward extrusion, ' the metal flow is opposite to the direction of the punch movement and a cup-like shape can be produced. In ' forward extrusion ' the metal flow is in the same direction as the punch travel and a shaft may be produced from a slug of greater diameter or a hollow shaft may be produced. It is also possible to have combinations of backward and forward extrusion at the same time and to include some form of cold-heading operation.

Typical examples of work produced in steel, so far, are cartridge cases, spark-plug bodies, gudgeon pins, prop-shaft ends, bearing races and valve tappets, shock absorber and other hydraulic cylinders. This is a most interesting and highly manipulative cold-forming process.

Impact extrusion is a process applied to non-ferrous metal blanks, usually from strip. The process is exactly as described under extrusion except that it has been found practical to employ only quick-acting mechanical presses, and a theory exists that the extreme degree of manipulation which is possible takes place in a fractional period of time during which the metal has insufficient time to manifest work hardness. It is the most popular method for cold-working brass and aluminium sheet and typical examples are radio-valve

cans, bicycle pumps, coffee percolator tubes, etc. Toothpaste and similar tubes are an even older application in the softer metals.

Drawing is a process of manipulating metal by pressing through a die and is more commonly applied to the fabrication by deep drawing or pressing of sheet and strip forms. However, it is a term and an operation sometimes borrowed for the manipulation of more solid material workpieces and is employed to effect a reduction of wall thickness in hollow components where, because of the length of hole involved, extrusion would be impracticable. Cold reduction is effected by tensile loading as against compressive loading in the other processes.

Coining and embossing To a certain extent these terms are synonymous when applied to the production by press working of coins, medals, plaques, key blanks, etc., from strip form and in most cases a relatively low deformation or change of shape takes place.

Coining is also used to describe a forging operation carried out mostly cold to impart closer tolerances, smoother surface finish or special markings on forgings which have been produced hot.

Cold forging is a term which has been somewhat loosely applied to some of the above processes and to indicate attempts made to work metal by forming or pressing it to a desired shape. It attempts to imitate hot forging, but is restricted by the design of the article and the limited plasticity of the metal. To produce in the cold condition any desired shape in steel and thereby to imitate hot forging is the object of present endeavours, but a more suitable term might be ' cold forming. '

Advantages compared with conventional forging

Material utilization efficiency A much more efficient material utilization is effected in the shape of the part, which does not adapt itself to the technique of die filling by flash formation and control. In making a rough-forged cold shape, the amount of any excrescence produced is limited to a burr or very light fin. In addition, it is usually not necessary to make allowance for rough machining as it is customary with drop forging technique. Cold deformation may in favourable cases offer up to 95% utilization of stock material, whereas this figure is much lower in forging practice.

Material improvement Useful mechanical properties may be introduced in the article by the process itself, *e.g.* hardening. This may save a subsequent heat-treatment operation and may advantageously enable the utilization of a basically softer or cheaper grade of steel. The process is essentially one of improving metal and in this respect it may utilize ower-cost raw materials than in an existing process, *e.g.* coiled wire may be used, which is basically cheaper than cold-drawn hexagon stock.

The process is such that the grain flow produced may be applied with equal and often superior advantage to that inherent in a hot-produced forging; this is especially the case as the article may be produced fairly close to form and little or no subsequent machining (which introduces stress-raisers by cutting into the flow lines) is necessary. From the service viewpoint, the fine finish obtained is an improvement to the fatigue life, and is a considerable improvement over finishes obtained by the usual machining methods. There is no extrusion of flash and disruption of grain flow with cold-deformed parts.

Machining requirements and dimensional tolerances Finishing much closer to required form is possible, thus tending to reduce machine operations and cutting tool requirements and their attendant adverse effects on production schedules and final tolerances. The amount of machining required on cold-deformed articles is usually restricted to drilling, threading or the production of undercuts. As the operation is performed cold, greater standards of accuracy are possible than with conventional hot forging. There is no distortion during cooling.

Production efficiency No heating requirements prior to deformation being necessary, there is no need to provide furnacing equipment or to overcome the attendant disadvantages of scaling and metal loss usual in heating for forging.

A high degree of skill is concentrated in the tooling design method, with the result that less skilled labour may be used for production operations. Being employed cold, it lends itself to greater productivity and simplification of the handling procedures.

Product development For the conventional drop forger, the process offers a field of development outside the conventional market embracing two distinct lines of manufacturing endeavour. Firstly, it can invade the field of the machinist on selected jobs, especially those of a hollow nature such as spark plug bodies and gudgeon pins, etc.; secondly, it may find application in selected cases for articles formerly produced by deep drawing or by the combination of deep drawing and metal fabrication, such as thin-walled tubes with heavy bases, flanges or bosses.

Disadvantages and product design restrictions

From the viewpoint of the conventional drop forger the cold process has the following disadvantages:

Materials It appears that a requirement will

exist for relatively clean steel in the fine-grained condition, with close dimensional tolerances and possibly also treated with some non-ageing additives.

Equipment Special equipment appears to be necessary for the efficient production of parts by cold deforming. Whilst the capital expenditure involved may be similar to that required for conventional forging, the equipment may only be used for cold deforming work. Such equipment appears to include robust mechanical crank and hydraulic presses, phosphate-coating equipment and possibly atmosphere-controlled annealing equipment.

Tooling The design and extent of the tooling required for cold deformation is such that some considerable degree of skill and tooling design is likely to be necessary and in addition the economics of tooling may be such that the total advantages to be gained by cold deforming are not justified by the economic production run available.

Product design Cold deforming imposes some restrictions on product design.

In general, product design for cold deforming advantageously follows a general shape of a symmetrical and circular nature. Change of section may be accommodated, depending on its extent. It appears that quite small fillet or corner radii are to some extent desirable features, but that some allowances should be considered, if possible, in respect of corner radii being not filled. Forging draft and extraction tapers are, in general, not required and are, in fact, a hindrance to the process. The general design should be such that ejector mechanism can be used on the job during production.

Where the part is cold headed, a ' head volume of stock diameter required to fill it ' ratio of $4\frac{1}{2}$ to 1 is regarded as the limit of endeavour with facility. In a hollow extruded component a penetration ratio of 5 to 1 dia., without inter-stage annealing, is regarded as the maximum limit of design in America, but a ratio of 2·5/3·0 o.d. is used elsewhere. The possible reduction in cross-sectional area on cold extrusion varies with design of the article, but it is not generally in excess of 30% without intermediate annealing. A design feature which is in most cases essential is that some surface of the component should be relatively free and not subject to full confinement in the die; this can often be arranged by the careful consideration of the component and the metal flow involved in production.

The surface finish obtained is of the order of 0·3–3·5μ, and is intermediate between values for finely ground and honed surfaces. The steels employed for cold deforming so far include plain carbon steels up to 0·3% carbon and include all the case hardening types, even those having some small alloying content, but greater difficulty is encountered with steels having a carbon content in excess of 0·2%. However, when considering steel specifications the possibilities of utilizing the work-hardening properties of the softer grade of steel could always be examined.

In the field of cold heading, the product design must be such that a substantially increased diameter or cross-sectional area is required somewhere along the length of the component, usually at one end, and such articles as standard and special fastenings, *e.g.* bolts, nuts, screws, plugs and bearing races, etc., are produced from coiled wire or rod, usually in progressive set-ups; such an arrangement entails a large volume to justify the tooling and set-up times.

In cold extrusion when using vertical press equipment, the product design is not quite so restricted, but usually embraces a hollow component or one having some boss feature in the design which serves the important function of a relatively free surface during extrusion.

Hollow components can be extruded with or without a bottom end, and with stepped internal and external diameters if these are progressive. Flanges or shoulders may be located on the component, but preferably at one end. Internal diameters of different dimensions should be blended or should have as large a radius as possible. One American classification on the size of extruded steel parts stipulates a weight range of $\frac{1}{4}$–15 lb., diameter 1–6 in. o.d., $\frac{1}{2}$–4 in. i.d., with lengths up to 20 in. Dimensional tolerances vary from 0·003–0·010 in dia. with a concentricity of 0·015 in T.I.R. and an eccentricity between different diameters of 0·010 in.; ovality of diameters may vary up to 0·6%.

Equipment, tooling and processing requirements

Cold-extrusion equipment The manipulative equipment for cold extrusion consists mainly of vertical presses, although horizontal toggle-action presses are available in Germany. The vertical press arrangement has the inherent facility of providing a location or natural fixture for the work-piece, but does call for the preparation of cut slugs and cannot use continuous raw material; hence automated feed may be necessary to obtain the maximum efficiency and overcome the work-handling problem.

Both mechanical-crank type and hydraulic presses have been used for cold extrusion and their relative advantages and disadvantages have given rise to controversy which in most cases has been settled by using the type of press already available or most easily procurable. Original German development was carried out on mechanical presses, but the weight of evidence in recent times would appear to favour hydraulic presses over long periods of use.

Presses originally designed for drawing and stamping do not lend themselves to cold extrusion as they lack the essential rigidity and stiffness in the bed, frame and mechanism, and are frequently under-powered. Most published information on presses designed for extrusion emphasizes the importance of great robustness of all members.

For mechanical-crank presses it must be said that they can usually be operated faster, they have more opportunity to deliver an impact blow where this is advantageous, as it may be possible in some cases, they are more easily mechanized and consume less power. Mechanical knuckle-joint presses have a load/stroke diagram which makes them more suitable for coining and flattening. Hydraulic presses lend themselves to stroke and speed control and are generally considered to be easier to maintain over a long period of time and are kinder to the tooling. The preference between the two kinds of presses can easily be vested in the number of strokes per minute against the capital expenditure involved.

Because the process is new, specifications for such presses are only just being developed, but it can be stated that, as well as having adequate power and stability of construction, a desirable press for cold extrusion should possess adequate stroke and daylight and have plenty of tooling space. It must be fitted with top and bottom ejectors and a minimum speed of 15 strokes/min. is desirable. The implications of working speed are not yet fully understood, but it appears that a speed of 12 in./sec. is not excessive.

Cold-extrusion tooling The high loads and stresses imposed on cold-extrusion tooling are such that every possible idea and attribute of tool engineering must be used to ensure success. The work-piece under deformation calls for minimum pressures of at least 120 ton/sq. in. and is considered to behave almost entirely as if under fluid pressure, so that such pressures must be withstood by the tooling without failure. The limitation of available tooling materials appears to be such that both punch and die are elastically strained during the operation, and in addition the reactant forces change considerably during the operation and are greatly affected by the tooling design and finish.

The amount of elastic deflection is such that die ejectors of substantial proportions are necessary to relieve the work-piece, and is also such that some not insignificant variations of dimensions from cold-die sizes can occur, and due allowance in die construction must be made.

There is a natural heating up during the work cycle to 200–250° C. (intrinsic surface temperatures are believed to be higher) and due contraction allowance—usually 0·0025/in.—must be made for this.

The punch is usually long and requires the utmost assistance from bolster and sleeves to maintain its dead axial alignment. Punch nose design is important and can affect grain flow in the work-piece and the loading and the productivity of the punch and its ability to retain the phosphate coating on the work-piece. Punch materials must be such as to give the highest possible resistance to compressive stress, fatigue and wear. Some investigations into aspects of warming the steel before extrusion indicate that the load on the tooling is reduced with material heated to approximately 300° C.

The tooling problem in regard to the female die is somewhat easier, as space exists to support the die itself with various combinations of liners, die pots and bolsters. The usual practice is to have three concentric rings, the centre one being the die proper, and each outer concentric ring is shrunk on so as to impose a compressive hoop stress on the inner ring, as by this means the tensile hoop stress on the bore of the die is kept down to suitable limits. Very careful machining, heat treating, sizing and shrink practices are required. Some tooling practice involves the use of a substantial pressure anvil to take the major load, and this may be used as an ejector.

Cold-extrusion processing The essential requirements of cold extrusion processing are accurate cutting of the slugs, lubrication and, if necessary, interstage annealing.

Cutting is usually performed on cold saws and is subject to weight checking. If shearing is adopted, a coining or preforming operation may be utilized to remove burrs and consolidate the dimensions.

If annealing is required to promote maximum plasticity in the processing, temperatures up to 750° C. are involved, usually with atmosphere control.

The most necessary processing requirement is that of lubrication, which takes the form of careful pickling and clearing and bonderizing or phosphating; as many as nine treatment stages may be involved which have for their object the promotion of a zinc phosphate coating keyed to the metal surface and impregnated with soap or other lubricant.

Conclusions

In concluding this brief review of developments in cold deforming, it is natural that those of us who are engaged in drop forging will be anxious to know what these developments mean in terms of both opportunity and competition.

Drop forgers cannot fail to be attracted by a process which saves metal and saves it cold, and at the same time produces an improved product, but it is obvious that the process places some fairly

close restrictions on the shapes that can be made. Components already included in the drop-forged inventory, which have suitable shape and dimensional characteristics and at the present time involve considerable machining, would be high on the list for consideration. This group might include a good many gears.

The author does not envisage the cold forging of components having such general shapes as crank shafts, connecting rods and steering arms, etc. The process is more limited by the high working stresses in the tools rather than by the ductility of the metal; and its use to produce shapes other than those mentioned will be governed by developments in tooling materials.

In the mass production of machined components by Bar-automatics, it is not unusual to waste 50–70% of the raw material in swarf, and it is in this field that cold forming can and is being developed to advantage, *e.g.* the average Steering tie rod ball stud is machined from 1¼-in.-dia. stock with about 70% scrap; at least ¾ lb. of good steel is wasted every time one of these is made. Another example is that of gudgeon pins formerly made from tube, but by extrusion a much smaller slug can be used.

One of the latest applications for cold extrusion is the production of stepped propellor shafts in stainless material for outboard motors. An installation in Milwaukee, Wisconsin, U.S.A., is reported to have about 100,000 such shafts roughly 12 by 1 in. dia. from one set of carbide dies with, of course, considerable economies over the former method of machining the steps.

A useful result emanating from the development in cold deforming is that some research has been instituted on the behaviour of metals under close deforma ion, a matter which is almost impractical to establish with hot working conditions. The National Engineering Laboratory in Britain is carrying out experiments on cold forming which include the influence of specimen geometry, and the effects of repeated impact, varying degrees of strain and exothermic effects on the deformation produced. Perhaps from this direction we shall ultimately arrive at a better understanding of some of our problems in hot forging.

Summing up these developments in cold forming, it does seem that the kernel of all the matters restricting its development is that of the available tooling materials. In any problem involving the shaping of metals, the stresses imposed in the work-piece must be subordinate to those in the tooling; and it does seem that to avoid the ancient and well-proven method of heating to promote plasticity is to invite trouble. If, one day, the forging process could be fully automated, the heating and forming of steel in vacuum might have a similar and possibly greater scope than that of cold forging.

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done in overcoming the inertia of the moving parts and clutch losses during engagement amount to 20% of the work done per press cycle, and failure to assess these factors leads to serious under-rating of the drive capacity; and (3) the graph of drive shaft flywheel torques shows abrupt peak loadings calling for increased drive motor capacity to avoid overheating. It is thus clear that calculation of drive motor capacity should be based on the mean square torque, and not on the mean torque.

Determination of the dynamic loads operating on moving-type cutting mechanisms of cold press machines. YU. A. MIROPOL'SKII. Pp. 27–31.
Miropol'skii presents a method of determining such loads on the actuating mechanisms of the cutting devices of such machines in relation to the optimum sinusoidal law of the change in the acceleration of the push rod.

The use of variable delivery pumps for the drive of hydraulic presses. B. V. ROZANOV and E. V. BUBENINA. Pp. 32–34.
The article shows the possible shortening of the time of the working stroke of a press through the use of such pumps with automatic variation according to the pressure in the system without altering the electric drive motor capacity.

Induction heating of tubular products from ferromagnetic materials. I. I. BEZRUCHKO. Pp. 35–39.
Bending, upsetting and press work on semi-finished tubular products often calls for local heating. An experimental study has been made of the distribution of temperature fields in tube walls, determination of the heating time, and determination of the electrical energy parameters of equipment for induction heating.

Experience in stamping elliptical covers of large dimensions. L. S. AMARYAN and B. L. MURAVICH. P. 40.

Forging turbine blades of large dimensions and complicated outline. P. M. PAVLOVICH. Pp. 41–42.

Titanium and methods of its mechanical working. I. I. KRYMSKII. Pp. 43–46.
A review of experience in the Western World.

The use of backing plates for the top forging tools on steam-hydraulic presses. V. G. MEDVINSKII. P. 47.

Russian forging journal

Abstracts from the Russian forging journal — Kuznechno - Shtampovochnoe Proizvodstvo, June, 1960, **2.** *This is the second year of this journal devoted specifically to forging. Abstracts of the more important articles are given in* METAL TREATMENT *each month.*

Determination of the force required for the compression of forgings. E. N. MOSHNIN and N. M. ZOLOTUKHIN. Pp. 1–5.

Investigations, in particular of the temperature distribution in the surface and central layers of large forgings (ingots up to about 120 tons), explain the reasons for the considerable fall in resistance to deformation during compression with the increase in the weight of the forging piece, and reveal the existence of a 'scale' factor during the hot mechanical working of metals. It is concluded that (1) during the forging of carbon and low-carbon steels on hydraulic presses the thermal effect due to plastic deformation does not materially affect the resistance to deformation; (2) the resistance to deformation during compression should be taken to be the value of the resistance to deformation relevant to the mean temperature in a cross-section of the forging; (3) the mean temperature is determined from the condition of distribution of the temperature in a transverse section of the forging in accordance with a parabolic relationship; (4) if data are used from mechanical tests on small specimens for the determination of the resistance of the metal of a forging to deformation, certain important corrections must be made; and (5) the real resistance to deformation during the compression of large forgings is 50–80% of the value determined during mechanical tests at a temperature corresponding to the mean temperature of the forging.

Deformations and stresses during extension. I. YA. TARNOVSKII, A. A. POZDEEV and S. G. PUCHKOV. Pp. 6–10.

In order to develop optimum forging processes, work was carried out to reveal the dangerous stage of forging, to investigate the state of stresses and deformations in relation to the basic parameters of the technical process, which minimize the formation of external and internal defects and promote the elimination of the voids existing before forging.

The formation of laps during the free upsetting of cylindrical forgings with a central aperture. A. V. ALTYKIS. Pp. 10–12.

The ratio $d : D$ is plotted against the ratio $H : D$, where d is the i.d., D the o.d. and H is the height of the cylindrical body with a central aperture. The plot based on the author's experimental data is compared with similar curves adduced by Stepanskii and Tarnovskii and Pozdeev. Forging pieces with sufficiently low dimensional ratios will be free from laps.

Determination of the plastic properties of sheet steel on the basis of hydrostatic testing. L. A. RUBENKOVA. Pp. 12–13.

A comparison is made between monoaxial and biaxial extension to test deep drawing properties. Experiments were carried out on 0·08% C rimming steel; they show the advantages of biaxial, hydraulic testing.

Design of lateral expansion cavities for forging dies in the stamping of axi-symmetrical components. A. Z. ZHURALEV. Pp. 14–17.

During the stamping of a large number of components in closed or semi-closed dies on horizontal forging presses, crank-drive presses and forging hammers it is necessary to draw off the excess volume of metal into lateral expansion cavities; the article sets out a method for their design calculation. Their use creates optimum force conditions during forging, so that standard metal consumptions may be decreased by 4–8% and the life of closed dies may be increased to that of open dies.

The resistance of the metal to deformation during hot forging in crank drive presses. V. S. GOFFENSHEFER. Pp. 18–22.

A study is made of the changes in the tensile strength of structural steels at the forging temperature in relation to the rate of deformation during forging, defined as the change in the degree of real deformation in unit time. Methods are given for calculation of the rate of deformation in relation to nominal press force, final forging temperatures of various sections of the forging piece, and the real resistance to deformation.

An experimental investigation of the electric drive of a hot forging press. I. A. REMIKH and N. I. VASIL'EV. Pp. 23–27.

In a series of instances the electric drive motor capacity and the amplitude of the flywheel torque are found to be inadequate. This is due to incomplete calculation of the losses in the kinematic links of hot forging presses and in the dynamic operation. The aim of the research carried out on a 630-tonne press was determination of these losses and derivation of a working graph of the loads on the flywheel shaft. It was found that (1) the idle running losses on the transmission are considerable and should be calculated when assessing electric drive motor capacity; (2) the work

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Bellows production at Edinburgh

Welding a bellows flange

BELLOWS FABRICATION is an industrial development which demands considerable research into processes and materials and also exhaustive testing of the product and of the materials used. At the Chesser tube works of Munro & Miller at Sighthill, Edinburgh, strong emphasis is laid upon these aspects in the production of an extensive range of bellows.

The bellows joint consists primarily of a length of corrugated tube, the corrugations being termed convolutions. These convolutions may be mounted in a number of ways according to the duty being performed. A simple axial bellows has flanges at each end of the convolutions and is secured simply to the adjoining pipe work. In other forms, however, the bellows may be gimbal mounted or may be provided with hinged restraints in order to absorb internal pressure loading. Articulated bellows, developed for duties such as in tanker loading lines, consist of two sets of convolutions separated by a length of tube and restrained by axial restraints.

Bellows are used primarily in pipe layouts to accommodate linear thermal expansion and contraction and they provide many advantages over both loops and sliding glands for this duty. As compared with loops, bellows joints accommodate greater movement and require fewer fittings and supports and their greater simplicity gives them superior flow characteristics. While sliding glands can accept a greater amount of axial movement, bellows joints can accommodate lateral and other out-of-balance movements and, in addition, provide a packless, leaktight joint. Absorption of vibration is also an important function of these units and they are widely employed with engine exhaust systems in addition to their more common use in pipe work subjected to conditions of varying temperature.

While the more standard-type bellows are fabricated in various stainless steels and are welded to mild-steel flanges, where corrosion is a particular consideration fabrication may be in Monel metal.

Development of bellows and suitable manufacturing methods entailed considerable research into suitable materials and heat treatment, as well as a considerable programme of experimental and theoretical stressing carried out in conjunction with the Heriot-Watt College.

Manufacturing programme

The sequence of manufacture starts with the selection of the required gauge and size of sheet required to form the necessary diameter, this being cut to size, rolled and longitudinally seam welded with an Argonarc automatic welding machine. The resulting tube is then inspected, blanked off, and the convolutions introduced hydraulically. By means of heat treatment before and after forming, the maximum ductility is maintained and the process itself exerts a uniformly rising pressure, resulting in steady, plastic flow of the material.

Although automatic welding is used wherever possible in joining of convolutions and flanges and other bellows components, nevertheless much hand welding is necessary, especially where the convolutions are in stainless steel and the flanges in mild steel. In such cases hand welding is carried out using Cooper & Turner Diadem zircon HR2 electrodes, the chemical composition of these electrodes providing a barrier which overcomes the tendency towards carbon migration from one steel to another and its precipitation as a non-metallic boundary on the edge of the weld. This electrode is designed primarily for welding fabrications in such heat-resisting steels as 25% chrome 20% nickel containing molybdenum, the Diadem variant zircon HR1 being employed for joining such steels as 23% chrome 11% nickel with tungsten to such lower-alloy steels as the 18/8 type. The weld is made in a single pass and, although the

continued on page 30

Metallurgy in nuclear power technology

5. *Fuel element canning materials—part III*

J. C. WRIGHT, B.Sc., PH.D., A.I.M.

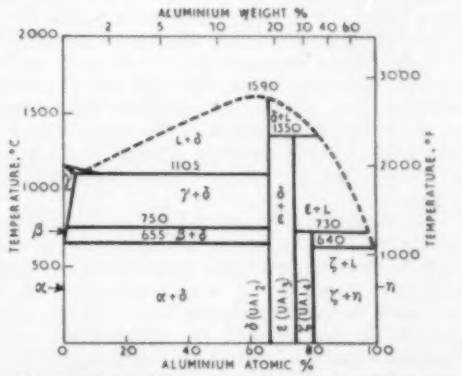
The metallurgy of nuclear power materials is developing on such a wide front and so rapidly that it is difficult for the non-specialist metallurgist to keep abreast with its scope. Dr. Wright, Reader in Industrial Metallurgy, College of Advanced Technology, Birmingham, outlines the subject in a series of articles which are appearing monthly in this journal

ALUMINIUM was the first metal to be used for sheathing uranium because its technology was well understood and it was commercially available. However, although its neutron absorption cross-section is low enough to be acceptable in thermal reactors, its strength at elevated temperatures is not very useful. This can be improved by alloying, but only at the expense of an increased neutron absorption cross-section. Thus, considerably improved creep resistance at 400°C. can be obtained with sintered aluminium powder (S.A.P.) and its associated oxide in a matrix of aluminium, or with precipitation hardened 6% copper-aluminium. When these alloys are judged on a basis of neutron capture cross-section per unit of creep strength they have little or no advantage over some usable magnesium alloys and certainly not over beryllium. This latter comparison, of course, ignores the cost.

Unless protected, aluminium reacts with uranium under reactor conditions. Above 200°C., intermetallic compounds form; first UAl_2, followed by UAl_3 and UAl_4 with increasing temperature and time (fig. 32).

The UAl_4 does not form an adherent layer on either Al or UAl_3. It cracks to a powder which gives a bad contact zone. Because of the high diffusion rate of aluminium into uranium, a mass transfer occurs and a porous zone then interrupts contact between the uranium and the aluminium. This leads to 'pimpling' of the surface of the canned element due to the increased volume of the compound compared with the reactants. This situation can be prevented by creating a barrier layer of suitable material between the fuel and the can which will remain stable with respect to both. Examples of workable solutions to pimpling are the use of an aluminium-silicon brazed bond which does not spread significantly with time in the neighbourhood of 300°C., and the provision of an anodic coating on the inside of the aluminium tubing. Anodizing yields an oxide layer which prevents reaction up to about 550°C.

Aluminium, in common with several possible canning materials, has a coefficient of thermal expansion greater than that of uranium (see Table 12). (Al 25 in./in./°C. × 10^{-6} and U 16–18 in./in./°C. × 10^{-6}). This situation can give rise to a thermal ratcheting process unless special precautions are taken. If an aluminium (or magnesium) canned bar of uranium is heated, the can expands more than the uranium. In a reactor operating with pressure on the elements, the can would tend to collapse on to the bar. During subsequent cooling the can would try to contract more than the bar and, being in contact with it,



32 *Constitutional diagram for the uranium-aluminium system* (USAEC *Report, BMI*/1,000)

the can material would tend to deform plastically. Continued thermal cycling produces permanent distortion and possibly failure of the can.

This defect can be prevented by one of two methods. In the first, the uranium can be a loose fit in a strong can which will not collapse and which would allow independent expansion and contraction of the fuel and the can. This method introduces difficulties in heat transfer and increases neutron absorption. The second method relies on holding the fuel and the can so closely together that they behave as a single unit. The bonding between the can and the fuel has been created by various methods, including high-temperature duplex rolling and extrusion, by powder metallurgy techniques and by casting the fuel into cans acting as moulds. In each case, continuity between the fuel and the can may be achieved, but it is necessary for the bond to be stable and for diffusion or reaction between the materials to be slow under reactor conditions.

The properties of aluminium as a sheathing material are not completely satisfactory and not apparently capable of development for much higher temperature applications than are in use at present. In alloyed form, it is still a possible material for use with high-pressure water reactors operating at surface temperatures up to 350°C.

Corrosion behaviour of aluminium

Pure aluminium is attacked by water or steam at temperatures about 150°C. The damage, which takes the form of serious blistering and exfoliation of the metal, is attributed to hydrogen ($2Al + 3H_2O \rightarrow Al_2O_3 + 3H_2$) penetrating the metal in the atomic form, combining into molecular hydrogen and nucleating bubbles of high-pressure gas in the aluminium and scale. Alloying can help to raise the working temperature to over 300°C. by additions such as 0·3–1% iron and 0·5–2·5% nickel, producing a range of materials known as the Aereal alloys. The alloying elements may promote recombination of hydrogen on areas rich in iron or nickel on the surface and thus avoid the diffusion of atomic hydrogen into the interior. Further improvement in the corrosion resistance of the Aereal alloys in steam up to 420°C. has been reported as the result of an addition of 0·05–0·1% of zirconium plus titanium.

The condition of the water has a strong influence on the rate of corrosion. The presence of salts such as chlorides and some metals such as copper stimulate corrosion of the aluminium. It is necessary to avoid using copper-containing components in contact with the water which could promote solution of copper and subsequent deposition on the aluminium alloy where galvanic action would promote pitting. Corrosion rates are also increased by increasing the velocity of the water in contact with the alloy surface, and by increasing the *p*H of the water. In this respect, aluminium alloys and mild steel in the same water circuit would together be incompatible with water because mild steel requires a high *p*H to inhibit its own corrosion.

In coolants other than water, aluminium responds variously. Its resistance to corrosion in air, or carbon dioxide, or helium is good and its life as a canning metal in these coolants would be limited by factors other than oxidation resistance. On the other hand, aluminium and its alloys are attacked by most of the possible liquid-metal coolants.

Fuel cans for low-temperature water-cooled reactors, including boiling water reactors, may be fabricated in aluminium alloys and give a serviceable corrosion resistance. At the temperatures involved in the pressurized water reactor with high heat extraction rates, aluminium alloys are not as satisfactory for long-term working as zirconium alloys. The choice between aluminium alloys and zirconium alloys under these circumstances is one of economics.

Magnesium

Magnesium has similar room temperature mechanical properties to those of aluminium, but is capable of slightly better elevated temperature properties under comparable conditions. Magnesium has a coefficient of thermal expansion (26×10^{-6} in. in.°C.) which is greater than that for uranium and, like aluminium, is susceptible to ratcheting. The neutron-absorption cross-section of magnesium (0·06 barn) is attractively lower than that for aluminium (0·22 barn). Neither the pure metal nor its usable alloys react with uranium in the solid state and magnesium and uranium are immiscible in the liquid state.

'Magnox' alloys

The desire to increase high-temperature strength, decrease high-temperature reactivity with the coolant circuit and develop good weldability without cracking led to the development of the oxidation-resistant magnesium alloys known as the Magnox series. These are based on magnesium with small amounts of beryllium, calcium and aluminium. Beryllium and calcium enhance the oxidation resistance of magnesium by the formation of a protective oxide film, but an addition of more than 0·2% of calcium severely impairs the weldability of the alloy. Beryllium is only slightly soluble in magnesium and cannot be added to any significant extent in the pure metal form. However, beryllium and aluminium form a eutectic system and both can be added at the same time as a 'hardener.' Beryllium has also been alloyed with magnesium,

by Russian workers, by co-distillating magnesium and beryllium, condensing *in vacuo* and homogenizing the deposit so formed.

The best known alloy of the Magnox series so far is Magnox A12, which contains 0·1% beryllium and 0·2% calcium. The alloy has given satisfactory service in gas-cooled reactors of the Calder Hall type. Extensive but successful developments were made in order to eliminate flux, oxide and beryllium-rich inclusions from Magnox, all of which adversely affect the corrosion resistance of the alloy.

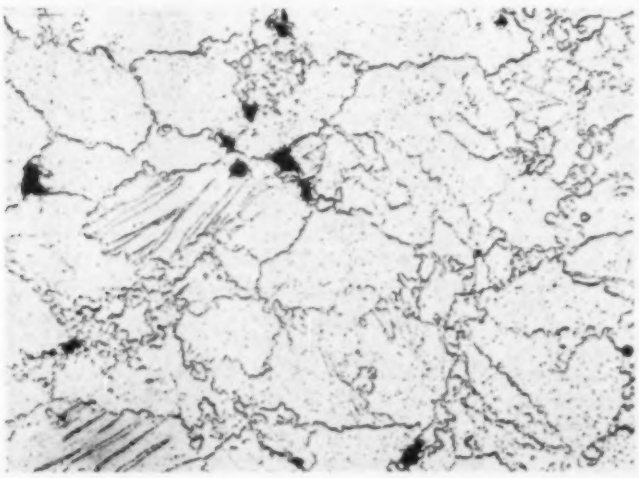
Heating Magnox A12 to temperatures above 400°C. leads to quite rapid grain growth. For short irradiations the properties of even relatively coarse-grained cans appear adequate and at working temperature there is no lack of ductility to accommodate any strains. But at positions in the reactor where the temperature is lower than the average the ductility of coarse-grained Magnox is not as good. A further disadvantage of grain coarsening is that intergranular voids, produced during creep of Magnox in service, are more prone to join up than in fine-grained stock. Although there is no evidence of failures due to these intergranular voids, it might become a limiting factor if a high performance was required.

The formation of these intergranular voids is known as cavitation and the number of cavities increases with decreasing strain rate for a given strain. At a given strain rate, the number and size of the cavities increases with total strain and they tend to form preferentially in grain boundaries normal to the stress axis. Fig. 33 illustrates cavities formed at 200°C. with a creep stress of 1·7 ton sq. in. It seems likely that, at working temperatures, the properties of Magnox A12 are not radically affected by irradiation since the service temperature is above the annealing temperature. The thermal conductivity of Magnox A12 (0·276 cal./sec./cm./°C.) is lower than that of pure magnesium (0·411 units) but is adequate, being comparable with beryllium and much better than that of zirconium.

Generally, the properties of Magnox A12 are satisfactory. Its limitations are (1) a rather low strength at temperatures over 450°C., when it may be imagined as forming a rather stiff plastic case around the fuel rod but unable to contribute to any further load bearing, and (2) the fact that magnesium melts at 650°C. and has an appreciable volatility even below its melting point. Thus, although there may still be scope for some slight improvements by alloying magnesium, development of high-temperature reactors requires alloys outside the magnesium systems.

Of other magnesium alloys, magnesium-zirconium alloys have comparable oxidation resistance, better fine-grain stability but inferior creep resistance to Magnox A12. The relatively strong magnesium/manganese alloys introduce the problem of the high neutron-capture cross-section of manganese and are not favoured. The magnesium/cerium alloys have poor oxidation resistance. Zinc and thorium have also been investigated as strengthening elements and, in parallel with S.A.P. aluminium, sintered magnesium powder is also of interest.

From a production point of view, magnesium alloys are more difficult to fabricate than aluminium alloys and, generally, extrusion is preferred to forging and rolling as the primary working process for magnesium alloy ingots. Techniques are now



33 *Cavitation formed in Magnox A12 at 200°C. with a creep stress of 1·7 ton/sq. in.*

available for rolling fins on to magnesium alloy cans or for impact extruding the multistart helically finned cans. The 'Integron' high-fin process* extrudes the fins from the outer wall of a tube by rotary forming. In this way, a fin height of up to $\frac{7}{16}$ in. can be produced on a tube of $1\frac{1}{8}$ in. dia. with a tube wall of as little as 0·06 in.

A mechanical method for locking Magnox cans on to uranium bars has been developed and can be extended to other canning materials. The uranium bars are provided with a series of circumferential grooves and, after canning, the composite element is heated under hydrostatic pressure so that the Magnox flows into the grooves. This helps to overcome ratcheting effects.

Mechanism of reactor corrosion

Corrosion of canning materials in both gas-cooled and water-cooled reactors usually results in a uniform attack on the metal to form an oxide layer. Pitting corrosion is rare in reactor technology. The uniform corrosion of most canning materials undergoes a marked acceleration after a time, depending on material and environment. This change is known as breakaway corrosion. The general case of corrosion rate can be expressed by the following function:

$$w = Kt^n$$

where w = weight gain in mg./cm.^{2}, t = time of exposure in days and n is a rate constant. K is also a rate constant, in fact that applying to the Arrhenius equation $K = Ae^{-Q/RT}$ where Q is the energy of activation, T the absolute temperature, R the gas constant and A is a rate modulus.

Referring to the equation $w = Kt^n$, when $n = \frac{1}{3}$, as happens in the initial stages of corrosion of many canning materials by air, CO_2, or water, the corrosion product is protective. Some time after the start of exposure, breakaway occurs and the value of n becomes approximately unity. The film of corrosion product on the metal becomes discontinuous and flakes away, becoming non-protective. The useful life of materials under these circumstances is virtually limited to their breakaway times, but for several materials there appears to be a critical temperature below which breakaway is not observed, even after extremely long exposures. Breakaway times are influenced by pressure and composition of the environment, composition and surface condition of the metal as well as by temperature.

The exact mechanism of breakaway is not settled, but it is believed that, in gaseous environments, the initial adherent film grows epitaxially with the metal lattice. At a certain critical thickness the resultant strain becomes so great that the protective film of corrosion product cracks and exposes fresh metal. Support for this theory is given by the fact that when thermal cycling is superimposed on the corrosion mechanism, breakaway occurs in shorter times than is the case for isothermal exposure.

Pure magnesium is subject to breakaway corrosion in carbon dioxide at relatively low temperatures but this is largely overcome in the development of Magnox A12. This alloy is satisfactory in CO_2 up to about 450°C. and further improvements in the alloy might lead to a maximum working temperature of 475°C. Beyond this temperature the alloy would probably burn in CO_2 and magnesium would also be lost by volatilization.

The protection gained by the dilute alloying of magnesium may be explained as follows.

The rate of corrosion of a metal depends on the rate at which the metal ions can diffuse out, and the corrodant ions can diffuse in, through the corrosion product film. This rate of diffusion depends, among other factors, on the energy required for an ion to squeeze through the field of the appropriate lattice. If, now, the metal being corroded is alloyed with an element which will go into solid solution but whose ions are slightly smaller than the parent lattice ions, the resultant corrosion product film should have a contracted lattice compared with the parent lattice. This is so because the alloying ion, being small, will tend to diffuse more easily to the surface and so the corrosion product film will be enriched in alloying element compared with the material being corroded. Because the alloying element ions are small, the corrosion product lattice will be contracted. The energy required of a parent metal ion or corrodant ion to squeeze through the corrosion product lattice is therefore increased and corrosion slows down.

Thus the formation of magnesium oxide on Magnox is limited. The beryllium ions, being smaller than magnesium, will diffuse to the outer surface more quickly and concentrate in the outer parts of the film. The film lattice will become compacted and make it very much more difficult for fresh magnesium ions to diffuse through it. Under equilibrium conditions, magnesium reacts with CO_2 as follows

$$Mg + CO_2 \rightarrow MgO + CO$$

but it has been established that amorphous carbon forms part of the scale when magnesium alloys react in CO_2. It appears then that the above reaction proceeds in two stages

$$2Mg + CO_2 \rightarrow 2MgO + C$$
$$C + CO_2 \rightarrow 2CO$$

but at temperatures below 600°C. the carbon

* I.C.I. trade name.

reacts only very slowly with CO_2 so the carbon produced in the first stage of the reaction is not entirely removed. Thus the weight increases measured experimentally when magnesium alloys react with CO_2 are not entirely due to oxide but to carbon as well.

Provided breakaway has not occurred, the oxide scale is firmly adherent to the metal and no intergranular penetration of the metal by oxide occurs. Slight pitting has been observed at and above 500°C. but this does not increase noticeably with time.

Pure magnesium reacts in water at temperatures above 70°C. and no satisfactory performance for water-reactor purposes has been achieved by alloying. Similarly, magnesium and its alloys would not be considered for use as a canning material in liquid-metal-cooled reactors.

The properties of zirconium

Zirconium exists in two allotropic modifications: the alpha phase, stable up to 865°C., is close-packed hexagonal; and the body-centred cubic beta phase is stable above this temperature. The metal has a relatively high melting point (1,850°C.), but its mechanical properties are not very encouraging at moderately high temperatures. The U.T.S. decreases from about 25 ton/sq. in. at room temperature, with ductility of about 25–30% elongation, to about 5 ton/sq. in. at 500°C. and, as might be expected, the creep resistance of zirconium at 400–500°C. is poor. At 400°C. a stress of a little less than 3 ton/sq. in. produces an average creep strain rate of 0·002% per hour over a period of 2,000 h. in pure zirconium. The pure metal is unlikely to be used in structural work at elevated temperatures, but some improvement may be expected by alloying, particularly with small percentages of molybdenum, tin and aluminium.

From several points of view, however, zirconium is an attractive reactor material. It is compatible with uranium at temperatures up to about 850°C. and yet it may be bonded direct to uranium and the bond is strong. Its neutron-capture cross-section of 0·18 barns makes it an attractive canning material and, although it is very expensive compared with aluminium and the Magnox alloys, it is cheaper than beryllium. Part of the high cost is associated with the difficulty of recovering scrap zirconium without recycling it to the extraction process.

Two impurities normally associated with zirconium have deleterious effects. Any hafnium contamination must be eliminated because of the very high neutron-capture cross-section of that element (the extraction processes described earlier takes this fact into account) and the presence of nitrogen impairs both the neutron absorption cross-section of zirconium and its corrosion resistance in some possible reactor coolants, particularly water. The addition of tin, iron and nickel as dilute alloying elements to zirconium renders it resistant to water at temperatures of the order of 300°C. and neutralizes the deleterious effect of nitrogen on corrosion resistance.

The alloys are very good for canning fuel for water-cooled and high-pressure high-temperature water reactors. They are not as good for use in gas-cooled reactors but alloys of zirconium with 1–2½% niobium; with 1% copper and 1½% molybdenum; are promising in CO_2 at 500°C. There does not seem to be much future in zirconium alloys for work in CO_2 at temperatures higher than 500°C. however, although they are used for fuel element support brackets in CO_2-cooled reactors. Breakaway corrosion of pure zirconium occurs after 1,500 h. at 400°C.

The reactivity of zirconium with other possible coolants varies. The reaction with hydrogen starts as low as 150°C. but, as might be expected, no reaction takes place with helium. Of the liquid-metal coolants, bismuth attacks above 600°C. and mercury is also reactive, but the reaction between zirconium and liquid sodium depends on the oxygen content of the sodium. When the oxygen content is less than 2 p.p.m., the sodium is harmless, but when the oxygen content is 20 p.p.m., embrittlement of the zirconium occurs above about 250°C.

The Zircaloys

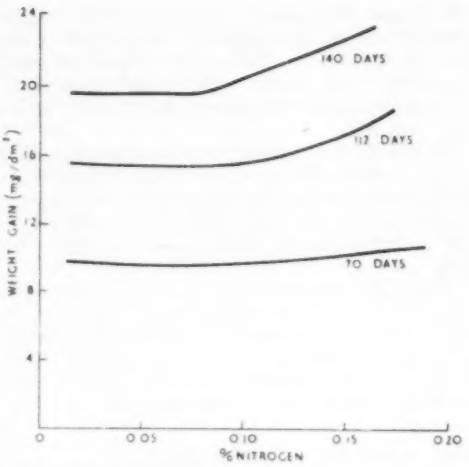
The best known reactor zirconium alloy is Zircaloy 2. The Zircaloys are a group of alloys containing from 0·5–5% tin. Zircaloy 2 carries 1·5% tin, 0·12% iron, 0·1% chromium, 0·05% nickel and a maximum of 100 p.p.m. of nitrogen, and its development has been able to make use of hafnium-free Kroll-process sponge zirconium. The tin forms a compound with the nitrogen preferentially and prevents it from impairing the corrosion resistance. (The nitrogen is known to accelerate the onset of breakaway corrosion.)

During the initial corrosion of zirconium in high-temperature water the corrosion film is a tightly adherent and protective black film. At breakaway, the corrosion rate suddenly increases, the film becoming white and flaking with increasing exposure. Hydrogen produced by the corrosion reaction appears to play a part in breaking down the film. Normally, pure zirconium does not exhibit breakaway characteristics at water temperatures up to temperatures of about 320°C. even after several hundred days' exposure, though at higher temperatures increasingly rapid onset of breakaway is observed.

Impurities, however, can cause early breakaway and nitrogen is particularly bad in this respect

(see fig. 34). The addition of tin to zirconium, as in Zircaloy, has the effect of increasing breakaway times and, in fact, when a transition does occur in the corrosion rate, the corrosion film remains adherent and consequently even the post-transition rate is less than it would otherwise be. It has been argued that the association between nitrogen and tin is explained by lattice defects. The nitrogen is presumed to be taken into the oxide film on oxygen sites as N^{3-} ions which, for electrical neutrality, means there will be oxygen-ion site vacancies. Such vacancies lead to increased oxygen-ion mobility in the scale and thus increased corrosion. When tin is present, however, it may enter the lattice as Sn^{3+} ions which neutralize the N^{3-} ions without vacancy generation. The functions of iron, nickel and chromium in Zircaloy 2 are not quite so well defined but they are believed to assist the formation of a stable oxide film. The film formed on Zircaloy is very protective and allows some variation in the flow rate, *p*H and impurity content in the water in contact with it.

The resistance of zirconium to corrosion by liquid metals is generally not very good, particularly at 600°C. and above. The liquid metals which are resisted best appear to be the sodium-potassium system.



34 *Effect of nitrogen content on the corrosion resistance of Zircaloy 2 (D. E. Thomas, Int. Conf.* PUAE, *P/537)*

Properties of beryllium

The most attractive property of beryllium from nuclear-power technology point of view is its low neutron-capture cross-section, 0·009 barns. This feature is so valuable that considerable efforts have been and are being made to overcome several detracting features of beryllium. The metal is very light, having a density of only 1·85 g. cm.^{3}, and possesses a very high elastic modulus above 40 × 10^6 lb./sq. in. Its melting point is moderately high at 1,283°C. The structure is close-packed hexagonal and the paucity of operable slip systems makes the metal very brittle. The crystals are also anisotropic and this combined with the brittleness makes mechanical working difficult. In addition the metal is toxic, and introduction of beryllium or its compounds into the human system can be extremely dangerous. The metal is also sensitive to contamination and a combination of these difficult factors has considerably hindered the commercial exploitation of beryllium.

Beryllium exhibits a wide range of mechanical properties dependent on the method of fabrication and, because of its anisotropy, on the direction and nature of the test. The best overall combination of mechanical properties is found in very fine-grained material as made by powder metallurgical techniques, but even then the pressed and sintered powder shows only about 2–4% elongation in tension. By creating preferred orientation, in sheet for instance, elongations of 30–40% may be developed in certain planes but not in all directions. Strength is in the region of 20 ton/sq. in. Cast beryllium has a tensile strength of about 10 ton/sq. in. and virtually no elongation when tested at room temperature.

A single crystal of beryllium slips most readily on the basal plane [0001] in the $11\bar{2}0$ direction. The next most acceptable slip planes are the prism planes $[10\bar{1}0]$, also in the $11\bar{2}0$ direction. Twinning occurs in tension on the $[10\bar{1}2]$ planes.

Randomly oriented polycrystalline beryllium cannot be extensively favourably oriented for deformation by slip, and therefore exhibits negligible deformation in tension at room temperature. Ease of fracture propagation increases with grain size, hence the brittleness of cast beryllium which is both random and coarse grained.

When both powder and cast billets are extruded into rods, the tensile properties in the direction of extrusion are markedly improved. After extrusion at 1,050°C., the U.T.S. of cast beryllium is increased to 20–25 ton/sq. in. and the elongation to 2–3%. Extruded powder compacts may give 40–50 ton/sq. in. tensile strength with 12–20% elongation. The improvement in properties is a result of preferred orientation produced by hot working. The texture is such that the basal [0001] planes are preferentially aligned parallel to the extrusion axis and with the $10\bar{1}0$ direction parallel to the extrusion axis. When a tensile force is now applied along the extrusion axis, slip on the basal plane is very limited, so that its low resolved shear stress does

not govern the overall properties of the bar. Instead, slip takes place mainly on the prism planes which, with a higher resolved shear stress, resist fracture and allow some extension. On the other hand, the transverse properties of extruded bars would not be expected to be good. The limited evidence at present suggests that transverse properties are poor, being of the order of 20 ton sq. in. U.T.S. with an elongation of less than 1%.

The same considerations apply to hot-rolled sheet, which may have an ultimate tensile strength of 40 ton sq. in., with an elongation of 10–15% in the direction of rolling. To achieve good ductility in all directions in the plane of the sheet it is necessary to cross-roll. If cross-rolling is carried out equally in both directions at 1,000°C., an elongation of 25% and U.T.S. of 30 ton sq. in. results in the plane of the sheet. Naturally, it is impossible to achieve favourable orientation in the thickness of the sheet at the same time as in the plane, and it is to be expected that strength and ductility in the direction normal to the plane of the sheet will be poor.

Tensile specimens prepared from cross-rolled sheet fracture after necking down in the plane of the sheet. There is no corresponding necking in the thickness of the sheet, indicating negligible ductility in this direction. This means that hot-worked beryllium will show reasonable ductility in the directions of working, but only if the stress system giving rise to the ductility is a simple one. With complex stress systems which rely to some extent on the unfavourable direction in worked beryllium, the metal may still exhibit brittleness. For instance, the type of stress system created in notched-bar impact testing would be expected to give inferior ductility and hence low impact energies to fracture. Generally, impact values for beryllium are, in fact, low but the exact effect of notches combined with preferentially oriented beryllium is not yet clear.

The use of beryllium as a canning material in advanced reactors requires an examination of the mechanical properties of beryllium at high temperatures, particularly in the range 400–800°C. Hot-extruded and warm-extruded powder beryllium show an almost uniform decrease in ultimate tensile strength as the testing temperature rises in short-time tests. Extruded cast beryllium decreases little in ultimate tensile strength until a temperature of 600°C. is exceeded and then the U.T.S. falls more sharply with further rises in temperature. The elongation of all the materials in the longitudinal direction increases to a maximum at temperatures between 200 and 400°C., but in the transverse direction little improvement is shown until a temperature of 550°C. is reached. The type of fracture which occurs in beryllium changes from transcrystalline to intercrystalline over the range 300–600°C. according to the method of fabrication. The strength of beryllium powder products is also dependent on grain size, being higher for fine-grained material at room temperature. On testing at increasing temperatures, however, the advantage of a fine grain size diminishes, until above about 600°C., for hot-pressed and extruded material, the coarse-grained material has the better properties.

The mechanical properties of beryllium at high temperatures are, as with other metals, very dependent on strain rate, but the amount of information freely available on beryllium diminishes as one progresses from hot-tensile tests through stress rupture tests to creep tests. It is evident that beryllium is not a highly creep-resistant material but it is not yet possible to say, on the basis of published data, to what extent its moderate creep resistance will limit its use in reactor technology.

All mechanical properties of beryllium vary considerably according to the method of production and are also very sensitive to the state of the surface and alignment of the test-pieces. It is essential in making critical tests to ensure that the surfaces of test-pieces are cleared of cracks by etching and possibly polishing. The influence of impurities on mechanical properties of beryllium is also very significant.

The influence of impurities on beryllium is the mainstay of one of the arguments explaining the room temperature brittleness of beryllium. Investigations into the influence of iron, carbon, nitrogen, hydrogen and other elements in trace quantities have been carried out with a view to explaining the brittleness on these grounds. Variations in the quantity of the trace elements individually and collectively have shown poor correlation with changes in ductility, and the purest beryllium produced to date has shown no appreciable ductility. However, the so-called purest beryllium still, as yet, contains sufficient oxygen (0·2 at%) to embrittle titanium and zirconium under equivalent conditions. Arguments for inherent brittleness irrespective of impurity content and based on physical properties have also been advanced.

Corrosion properties of beryllium

Beryllium oxidizes slowly in air at low temperatures, forming a protective oxide film. Temper colours appear just above 600°C. and the oxidation rate becomes significant above 800°C. until at 1,000°C. a white film is quickly formed. Beryllium is less reactive with nitrogen and beryllium nitride formation is not significant below 1,000°C., in air. Similarly, negligible reaction occurs with hydrogen below about 1,100°C.

In dry CO_2 at atmospheric pressure the attack

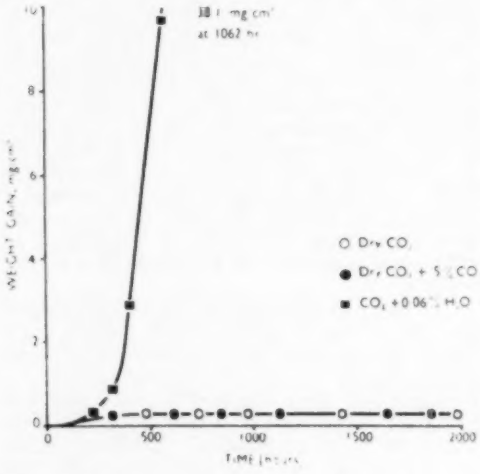
is negligible below 550°C. and the thin film formed is protective following a parabolic rate growth law. It is probable that the reaction takes place in two stages as for magnesium

$$2Be + CO_2 \rightarrow 2BeO + C$$
$$C + CO_2 \rightarrow 2CO$$

It is also possible that beryllium carbide may result and, in the presence of water vapour, beryllium hydroxide. At 650°C. (see fig. 35), the film becomes non-protective and breakaway dry corrosion is observed in the presence of about 0·05% moisture which is about the lowest practical limit for moisture in a reactor gas coolant circuit. Slightly higher moisture contents lower the breakaway temperature to about 600°C. The resistance of beryllium to purified helium is very good.

The corrosion of high-grade beryllium in pure water at temperatures below 100°C. is satisfactorily low, but some dissolved ions such as chloride, sulphate, cupric and ferric ions have a deleterious effect, generally giving rise to pitting corrosion. Gross contamination of the beryllium, particularly by carbide, or fluxes from the reduction process, is also detrimental to corrosion resistance. Generally, impurities in either the beryllium or the water become more and more detrimental to the corrosion resistance as the temperature of testing in water rises.

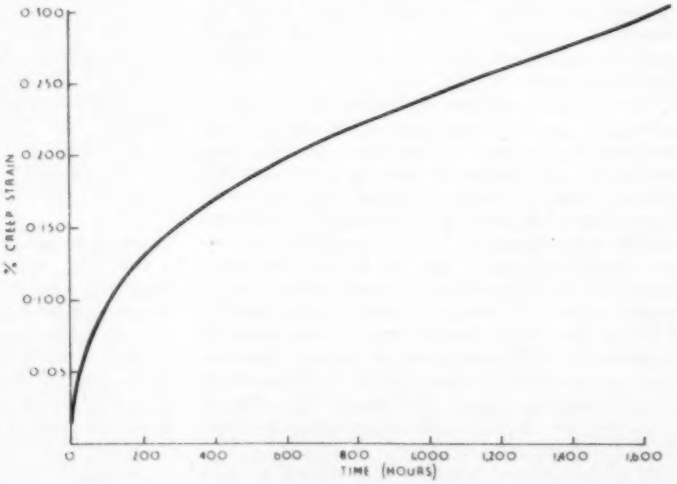
The performance of beryllium in liquid metals is somewhat variable, being particularly sensitive to the oxygen content of the liquid metal. This is because beryllium oxide has a high energy of formation and therefore beryllium may act as a 'getter,' becoming oxidized in the process. Assuming the oxygen content of the liquid coolant to be sufficiently low, beryllium appears to resist attack by sodium at 500°C., but under conditions likely to operate in power reactors the liquid-metal resistance of beryllium is rather uncertain at present.



35 ABOVE
Breakaway corrosion of beryllium
(*G.E.C. At. Energy Rev.*, **2**, (1), 14)

Toxicity of beryllium

In any discussion on the properties of beryllium it is necessary to survey its toxic effects because these govern the difficulty of handling the metal. The metal and its compounds are most dangerous when inhaled. With a heavy exposure, the maximum recommended limit being only 25×10^{-6} g. m.^{3}, an acute form of pneumonia can develop relatively quickly. Under continuous exposure the



36 RIGHT
Creep strain in niobium at 400°C.
(*J. Inst. Met., July*, 1959)

average concentration throughout an 8-h. working day should be less than 2×10^{-6} g./m.^{3}, and this is aimed at limiting the chronic poisoning effect of beryllium, berylliosis. The susceptibility of different people to the disease appears to vary considerably and there is no test for estimating this susceptibility. The disease can develop after periods of up to 15 years from the last time of contact with the metal. Finally, beryllium introduced into the body through open wounds produces ulceration, dermatitis and conjunctivitis effects.

To avoid these potentially dangerous toxicity problems it is necessary to take adequate safety precautions when dealing with beryllium, particularly when it is in powdered form. The greatest degree of protection is provided by glove boxes, but this considerably complicates manipulation. In most work it is possible to carry out work on powdered beryllium in glass-fronted cupboards equipped with powerful extraction fans so that dust is drawn away to a cleaning plant where it can be dealt with in bulk. Since beryllium is melted and cast *in vacuo* these operations are automatically sealed and fabrication of bare beryllium is safe up to about 600°C. provided no dust is created. At higher temperatures, sheathed beryllium working is also safe.

In machining beryllium, precautions are required. When a coolant is used, most fine particles and swarf are flooded away by the coolant and chemical treatments are necessary to recover the beryllium scrap from the coolant. Any beryllium which escapes the coolant will be air-borne but may be extracted with moderate air-flows. When no coolant is used, the rate of air extraction over the machining zone has to be high and, in cases where fume or fine powder is excessive, a glove box must be used.

Regular workers with beryllium normally need a cleansing zone between the beryllium handling area and the outside of the plant so that showers and a complete change of clothing may be used In addition, regular air sampling and analysis for beryllium contamination is necessary in the beryllium plant generally and also at specific high concentration points. Finally, although beryllium powder is not pyrophoric, except possibly in ultrafine sizes, beryllium in a fire could be dangerous and extra precautions are required in fire-proofing.

It is unlikely that Magnox alloys will be used in gas-cooled reactors at temperatures in excess of 475°C. For more advanced gas-cooled reactors where fuel-element temperatures of 600°C. or higher are envisaged, Magnox alloys definitely cannot be used. Beryllium is one of the most attractive metals for this field, in coolants such as dry CO_2 or helium. The ductility of beryllium increases markedly above 200°C. and, since it will be above this temperature in a reactor, the cold brittleness of beryllium is not likely to be an objectionable factor once the sheath is fabricated.

The major difficulties lie in production on a large scale and in economical fabrication. Work on conventional canning methods using beryllium is hampered by its low transverse ductility and, in addition, by the difficulty of obtaining ductile welds. Welding or brazing is possible, though complicated by the rapid oxidation of the metal in a normal atmosphere. Alternative methods of sheathing uranium, such as co-extrusion of uranium and beryllium in a similar way to be described later for sheathing powder compacts, have been tried and may provide a further solution to the brittleness problem.

Since beryllium has a poor resistance to high-temperature water and liquid metals under workable reactor conditions, it is not likely to be used for these types.

Properties of niobium

The physical and mechanical properties of niobium are not unique but, having a moderately low neutron-capture cross-section (1·1 barns) and very attractive corrosion resistance, the metal is of interest in reactor technology. Niobium has a melting point around 2,460°C. but no phase transformations in the solid state, being body-centred cubic in structure up to its melting point. Niobium in the annealed condition is soft and ductile, is easily worked and work-hardens less readily than most other metals. The mechanical properties, especially ductility, are very dependent on the impurity content, particularly the oxygen content. The influence of the oxygen content is illustrated in Table 13. A limited amount of creep data is available on niobium which confirms early assumptions that primary creep is rapid followed by long periods at low creep rates. Fig. 36 shows typical

TABLE 13 *Effect of oxygen on tensile properties of niobium (C. R. Tottle, J. Inst. Met., April, 1957, 377)*

Oxygen content, wt.%	V.P.N.	U.T.S., ton/sq. in.	Elongation, %
0·03	87	18·35	29·3
0·161	194	34·1	16·9
0·208	208	40·7	17·66
0·279	248	44·3	20·7
0·315	278	60·9	20·5
0·371	314	61·0	10·4
0·410	331	58·6	9·8

creep properties of niobium at a temperature of 400°C. and under a stress of 8 ton/sq. in.

Oxidation and corrosion behaviour of niobium

At room temperature niobium is resistant to oxidation, and only temper films form at 200°C. with no oxygen penetration or embrittlement of the metal. At higher temperatures, oxidation rates become increasingly serious and, even in CO_2 at temperatures of the order of 500°C. and above, linear oxidation rate laws are followed.

Niobium has excellent corrosion resistance in most acids, probably due to the existence of a thin protective oxide layer. The exception to the rule of good acid resistance is that solutions containing fluoride ions rapidly attack niobium, probably with the formation of complex fluoniobate ions. The metal is not very resistant to caustic alkalis, probably giving solutions containing the niobate ion.

A number of pure liquid metals do not attack niobium at probable service temperatures. These include lithium, sodium, mercury, tin, bismuth, lead and sodium-potassium alloy. These desirable properties result from the almost negligible mutual solubilities of niobium with any of these liquid metals, even at 600°C., and in some cases at higher temperatures. The presence of a small concentration of oxygen in the liquid metal causes a significant increase in the corrosion rate and corrosion may proceed by the formation of a non-adherent scale. However, its promising resistance to attack by liquid metals makes niobium an attractive proposition in liquid - metal - cooled reactors, particularly of the fast neutron type.

Other possible canning metals

Many more structural and cladding materials become available when a higher neutron-capture cross-section can be tolerated such as in a power reactor using enriched fuel or a fast neutron system. The range of materials acceptable in thermal neutron systems is widened if the canning is sufficiently thin. Thus, materials such as certain stainless steels, tantalum, molybdenum, vanadium and tungsten are of interest.

When using stainless steel, a reaction between uranium and iron, which can lead to a eutectic melting at 716°C., must be avoided. A brittle intermetallic compound also results from inter-diffusion of iron and uranium at a rate which depends mainly on the intimacy of contact and the temperature. The oxide film which is normally present on uranium is useful in preventing very intimate contact between the fuel and the can. The fuel might also be separated from the stainless steel can by a liquid sodium potassium layer which again prevents intimate contact but at the same time acts as a heat-exchange medium.

Some reactor corrosion problems would be overcome if materials such as niobium, molybdenum, tungsten, tantalum and vanadium could be used. They tend to resist attack by sodium monoxide in sodium potassium alloy better than most metals and have generally good high-temperature strengths. These new metals bring their own difficulties. For instance, molybdenum oxidizes very rapidly in air or CO_2 and is difficult to weld. Tantalum and vanadium are produced by small-scale powder metallurgy methods and are relatively scarce and costly, although tantalum is weldable. Both tantalum and tungsten have thermal neutron-capture cross-sections of the order of 20 barns and consequently could probably only be used in fast neutron reactors.

Bellows production at Edinburgh

concluded from page 20

welds are subject to radiography, no metallurgical failure has been experienced with this electrode during the past four years.

Post-weld treatment includes the grinding and polishing of welded surfaces with consolidated pneumatic Hicycle high-frequency electric grinders.

For some requirements, as for example, those of the nuclear power industry, semi-clean conditioning is carried out with Hicycle grinders fitted with Gem stainless-steel brushes. After further inspection, the bellows are treated with 'Trilac' to remove any final traces of foreign matter and are sealed in polyethylene bags.

The bellows are subjected to hydraulic and radiographic tests. The radiographic section is equipped with a laboratory in which are housed a caesium source and a Fedrex 150-kV. gas-insulated set, together with developing and illuminated viewing facilities. A mobile Pantak 150-kV. set is also available.

A typical example of the radiography carried out is the examination of a bellows of CrMo steel, welded to 1½ in. thick mild-steel flanges. For this examination the caesium source is employed, the exposure time being 12 hours and the source film distance being 18 in. The film used is Kodak M ultra-fine-grain film to obtain the highest possible sensitivity. At one end of the bellows, the thickness was 1½ in. and here lead screens were employed, these being omitted on the other 1-in. weld. The perfection of a development technique to suit these conditions where the same grade film is used for one exposure of two welds of differing thicknesses has been particularly successful.

This exposure time of 12 h. is convenient for overnight shots, but where quicker results are necessary, lead screens are employed throughout and the developing time is varied accordingly.

Effect of carbide stringers on the distortion of die steels during heat treatment

K. SACHS, PH.D., M.Sc., A.I.M.

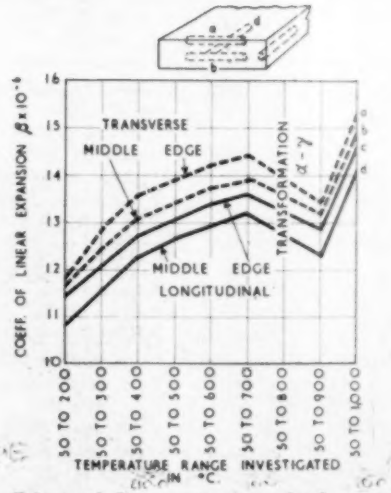
The causes and mechanism of distortion of die steels during heat treatment, the influence of the structure of the steel and in particular the part played by carbide stringers, are studied. The author is Head of Research Metallurgy Section, G.K.N. Group Research Laboratory, Wolverhampton, and his article will be continued in future issues

continued from last month

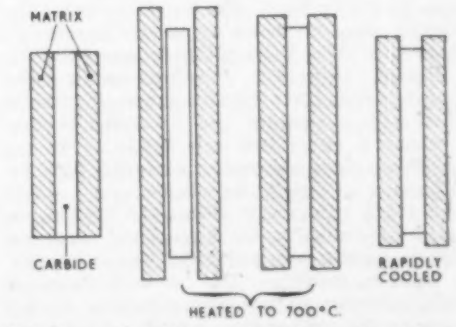
THE PRESENCE of carbides has been shown[16] to lower the coefficient of expansion of steel. This implies that directional alignment of carbides should lead to anisotropic expansion effects. This was confirmed by determining the linear coefficients of expansion of specimens taken longitudinally and transversely from core and surface positions of flat forged bar. Fig. 22 shows that the coefficient of expansion was less parallel to the longitudinal axis than in a transverse direction and less in the centre than at the surface. Thus specimens with coarse longitudinal carbides show the least expansion, those with fine transverse carbides most.

Evidently, the lower the coefficient of expansion in any direction, the greater the change of dimension on hardening in that direction. The reason for this is illustrated in fig. 23, which indicates the response to thermal expansion and contraction of a sandwich consisting of a carbide platelet between two plates of matrix material. If they were separated, they would expand on heating according to their respective expansion coefficients, the carbide about 30% less than the matrix. Since the phases are firmly bonded together, the carbide inhibits the expansion of the matrix material, although some mutual displacement at the interface will occur. This process is reversed in cooling, and the inhibition of contraction is expected to be more pronounced the faster the cooling rate.

A cycle of heating and cooling is likely to lead to an overall increase in the length of matrix particles.



22 *Coefficients of linear expansion related to carbide distribution in longitudinal and transverse specimens from a flat bar (J. Frehser[15])*
(*a*) fine transverse carbides, (*b*) coarse transverse, (*c*) fine longitudinal, (*d*) coarse longitudinal



23 *Influence of carbides on thermal expansion (schematic) (J. Frehser[15])*

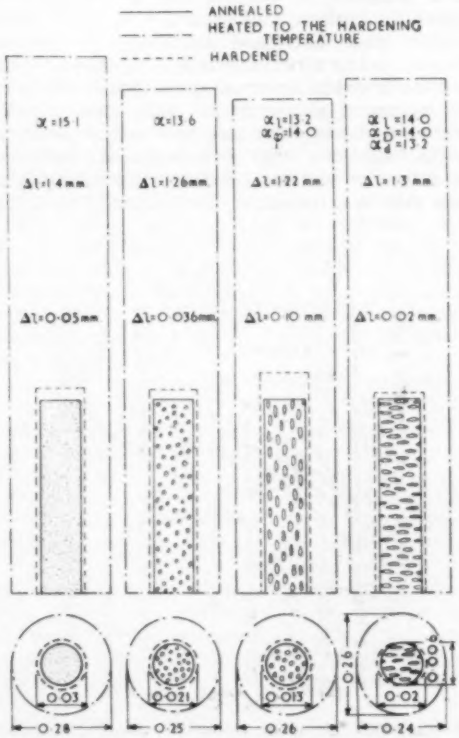
The carbides act like spikes, offering some resistance to the expansion of the matrix during the relatively slow heating cycle and exerting a much stronger inhibiting effect on contraction during a quench.

When phase changes and the associated dilatations of the matrix are superimposed on the dimensional changes due to temperature variations, the influence of carbide distribution on dimensional anisotropy becomes even more complex. The situation arising in the hardening of high-carbon high-chromium steels with plenty of carbide particles embedded in the matrix can best be explained with the aid of fig. 24. Dimensional changes are drawn to a greatly magnified scale for cylindrical bars 20 mm. dia. and 100 mm. long. The dimensions of the original specimens, in the annealed condition, are drawn in full lines. Expansion due to slow heating to the hardening temperature is shown by chain-dotted lines.

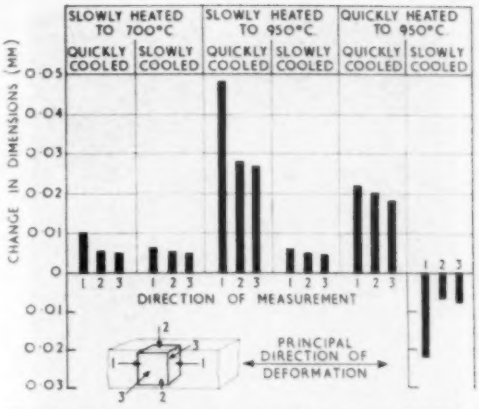
The left-hand diagram applies to a slightly hyper-eutectoid low-alloy steel with a coefficient of thermal expansion of $\alpha = 15{\cdot}1 \times 10^{-6}$ and shows isotropic expansion. A high-carbon high-chromium steel with uniformly distributed spherical carbides —if a steel like this could be made—would be expected to have a somewhat lower coefficient of expansion, $\alpha = 13{\cdot}6 \times 10^{-6}$, and it would expand rather less, though still uniformly in all directions, as illustrated in the second diagram. If the carbides are aligned in stringers parallel to the longitudinal axis of the cylindrical bar, they inhibit thermal expansion in this direction and this is compensated for by increased expansion of the diameter as shown in the third diagram. The fourth sketch refers to the same steel with the carbide stringers at right-angles to the longitudinal axis of the cylinder, and here expansion has been restricted in one transverse direction, so that there is increased expansion of the length and of the transverse direction normal to the carbide stringers, resulting in an oval cross-section.

The effect of rapid quenching for hardening is shown by broken lines. The low-alloy steel in the left-hand diagram shows an overall increase in volume of 0·35%; its length has increased 0·05%, its diameter by 0·15%. The high-carbon high-chromium steel has a higher content of carbides and of residual austenite, and its overall increase in volume is rather less, only 0·25%. For the case of spherical, uniformly distributed carbides, the increase of length and diameter is 0·036% and 0·105% respectively. When the carbides are aligned longitudinally, the longitudinal expansion has increased to 0·10% and the increase in diameter has fallen to 0·065%. The bar with transverse carbide stringers shows most expansion in this direction; the cross-section is oval, but its major axis is parallel to the carbide stringers, that is at right-angles to the major axis of the oval section at the hardening temperature. In a direction normal to the carbides, the dimension of the hardened bar is similar to the diameter of the original specimen.

It has been asserted that carbide inhibits the dilatations of the matrix in which it is embedded, that this inhibition operates most strongly at the interface, so that it is particularly strong in the direction of the carbide stringers, and that the inhibition is particularly effective when the temperature change is rapid. The experimental evidence of the distortion of steel containing elongated particles is consistent with these assumptions. If cohesion across the carbide/matrix interface inhibits dilatations of the matrix, the resultant stress will be relaxed by shear displacement along the interface under conditions resembling creep; the more opportunity there is for a creep mechanism to operate, that is the more time is available for creep deformation, the greater the dilatation that will actually take place. Thus slow rates of heating or cooling should give greater



24 *Mechanism of anisotropic distortion. The dimensions indicate the changes relative to an initial diameter of* 20 *mm. and length of* 100 *mm.* (*J. Frehser*[15])



25 Distortion experiments on cubes, confirming the inhibition of dilatations by carbides (J. Frehser[15])

changes of dimension parallel to the carbide stringers than rapid temperature variations. It follows that the anisotropic distortion of steels with carbide stringers should be much greater when they are heated slowly and then quenched or when they are heated rapidly and then cooled slowly, than they are likely to be when heating and cooling rates are similar.

Frehser confirmed this experimentally.[15] Cubes were cut from the steel with the direction of the carbide stringers parallel to one axis; they were heated and cooled at different rates and the bellying in or out of the faces was measured with a micrometer. The results are given in fig. 25.

The effect of thermal dilatations alone was studied by heating the specimens to a temperature at which no austenite was expected to form, 700°C. Slow heating to 700°C. followed by rapid cooling allows some expansion to occur but inhibits contraction more strongly, so that the specimen is elongated in the direction of the carbides. In slow heating followed by slow cooling there is, as predicted, no preferential directional effect on the dimensional changes.

When the experiments were repeated at a temperature of 750°C. the effect of transformation could also be investigated. Slow heating followed by quenching leads to a greatly increased expansion in the direction of the carbide stringers. The reverse procedure, rapid heating and slow cooling, leads to an overall contraction, which is particularly pronounced in the direction of the carbides, because the latter had a greater inhibiting effect on the expansion during rapid heating than on the contraction during slow cooling.

The experiments confirm that the reason for anisotropic distortion of steels with a skeleton of elongated carbides is to be found in the differential thermal expansion characteristics of the carbides and the matrix.

Warping due to variations in carbide structure

Frehser's experiments, described in some detail in the previous section, revealed in passing that coarse carbide stringers lead to sharper dimensional changes than finer carbides (Table 3, December, p. 491). In forged bars of fairly heavy section, from which tools and dies are usually machined, the carbides are normally much coarser in the centre than at the outside. This is due partly to more rapid cooling of the outside of the cast ingot and partly to somewhat heavier working of the surface layers in forging. Typical carbide distributions, and the influence of forging variables, are considered in a later section.

In the heat treatment of the die, adjacent layers of different carbide distribution (figs. 16–18, December) undergo essentially similar temperature cycles to which their response is slightly different. In general, a hardening treatment will lead to overall expansion, much accentuated in the direction of the carbide stringers, much less pronounced at right angles. This anisotropic enhancement of the expansion in the direction of the carbides is greater in zones where the stringers are coarse than in areas where the carbide particles are more or less rounded and the directional character of the structure is established by the arrangement rather than the shape of the carbide particles. Such differences in expansion characteristics of adjacent layers are likely to cause throwing and warping of the die.

Distortion and warping under these conditions has been proved experimentally by the development of a special distortion test.[18, 20] The requirements for a suitable test were laid down as follows:

1. The specimen should have a sharp change in section to make it particularly susceptible to warping due to asymmetrical martensite formation.

2. The specimen had to be cut from forged bar in such a way that different parts of the cross-section came from layers of obviously different carbide structure; this relation between structure and specimen shape had to remain constant at all points along the specimen. It followed that the specimen had to be long, thin, and parallel to the flow of the carbides.

3. The design of the specimen had to facilitate easy measurement of distortion and easy correction by grinding; this allowed the same specimen to be used many times.

Several sizes of forged bar were studied. The sharpest difference in carbide distribution, as revealed in the macrostructure of the cross-section, was found in the largest section examined, 2½ in. square. The effect of forging had penetrated to a

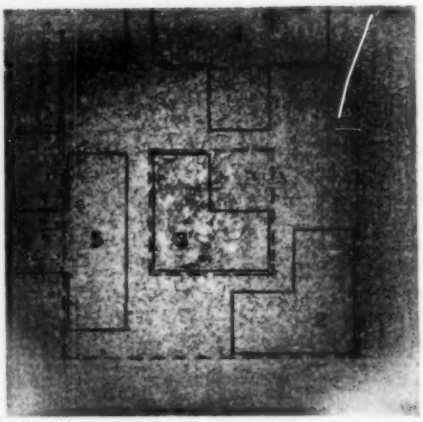
sufficient depth to draw out all the carbides into stringers parallel to the direction of flow, but the cross-section revealed many dendrites the original orientation of which was presumably in such a direction that they did not break up considerably under the forge hammer. The etched cross-section (fig. 26) shows an outer layer, about $\frac{3}{8}$ in. thick, with no clearly defined macrostructure; there is a central zone, $\frac{3}{4}$-in. square, containing fairly heavy carbide segregates; remnants of the dendritic structure are found in the intermediate layer.

Analysis samples were taken from the three zones, the amount of austenite retained after oil quenching from 960°C. was estimated from X-ray glancing-angle exposures at suitable points on the cross-section, and the amount of chromium carbide was estimated from the analysis figures and the ternary diagram.[2, 17] The results[18] are summarized in Table 4.

The variations in composition and constitution are very slight compared with the evident differences in structure, and distortion effects can be unhesitatingly attributed to the latter.

The distortion specimens were 8 in. long;[20] their cross-sections and their location in relation to the structure of the forged bar are shown in fig. 26. Specimens 1, 2 and 3 had a $\frac{3}{4}$-in. angle section with a $\frac{3}{8}$-in. web and were cut from the three separate zones. The other specimens had a T-section, $1\frac{1}{8}$ in. wide with a $\frac{3}{4}$-in. web; in specimen 4 the heavy cross member of the T was placed in the outer layer and the web in the intermediate zone, while the position of specimen 5 was reversed. The distortion characteristics of the different zones could be studied separately by means of the angle sections and the T-sections served to show the effect of the interaction of two zones.

The specimens were prepared with accurately ground parallel faces on both sides of the angle section and the corresponding webs, and on the wide face of the T-section and the bottom of its web. After heat treatment each ground face was placed on a standard flat marking-off plate and the parallel surface was surveyed with a dial gauge mounted on a Vernier height gauge. Readings were taken at points spaced 1 in. apart along both long edges of each face (in the case of the broad face of the T-section a set of readings was also taken along the centre line). Distortion of the specimen expressed itself in longitudinal bending and the maximum deflection was read off an auxiliary straight line joining the ends of the specimen, deflection towards the heavy section being recorded as positive, deflection towards the web negative. Only the readings on the face with a 'convex' bend were used for calculating the maximum deflection because only in this position was the specimen seated positively on the flat base without danger of rocking. Heat-treated specimens were pickled in 25% sulphuric acid, inhibited with a little glue, and the recorded deflections are not affected by surface unevenness due to scale.



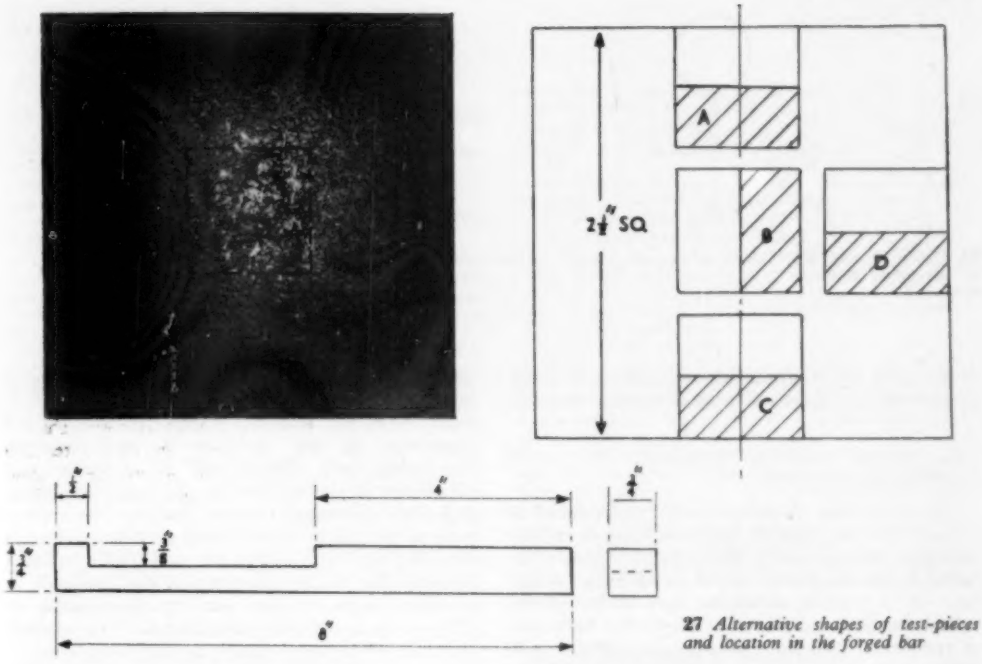
26 *Cross-section of 2½ in. sq. forged bar and disposition of distortion specimens*

Distortion measurements after annealing and after four hardening treatments are presented in Table 5 in which the maximum deflections on each specimen are recorded. It is interesting to observe that even annealing leads to some distortion, which is consistently opposite in sign to that produced by hardening.

The distortion of the T-sections, specimens 4 and 5, illustrates the effect of different expansion characteristics on the warping of components in heat treatment. In the T-section the thin web presumably cools a little faster than the heavier

TABLE 4 *Analysis results of samples taken from three zones of carbide distribution* (%)

Position	C	Cr	Retained austenite in matrix (X-ray)	Carbide in steel (equilibrium at 960°C.)	Martensite in steel	Retained austenite in steel
Surface	2·15	12·66	~ 12	12·0	77·4	10·6
Intermediate layer..	2·16	12·78	~ 11	12·2	88·1	9·7
Core	2·19	12·80	9	12·4	88·7	8·9



27 *Alternative shapes of test-pieces and location in the forged bar*

cross member, reaches the transformation temperature earlier and starts to transform to martensite earlier; however, this transformation can hardly have progressed very far before even the hottest part of the specimen reaches the M_s point and also begins to form martensite. Nevertheless the soft and ductile austenite may be expected to deform plastically as a result of the expansion of adjacent layers; when this deformed metal transforms the adjacent layers will be hard and brittle martensite which would require a much higher stress for plastic deformation. Thus, the stresses produced by the expansion of the last portion to transform are locked in the specimen and cause it to bend in such a direction that the heaviest part of the section is convex, *i.e.* a positive deflection.

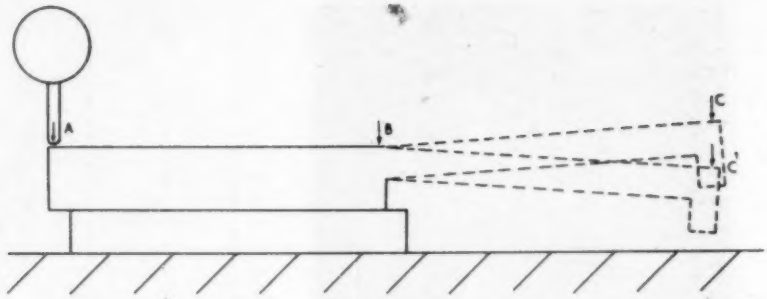
In specimen 4, the cross member was taken from the surface layer, the web from the intermediate zone. Hardening produced a negative deflection which implies that the web had expanded more than the heavier part of the section. Since the geometry of the specimens favours a positive deflection, the difference in expansion characteristics must be considerable. Specimen 5 showed a positive deflection, and it will be noted that the part which had expanded most, in this case the cross-member, was again taken from the intermediate zone. It can be concluded, therefore, that the zone between the heavily worked surface layer and the central segregate shows more longitudinal expansion than the surface layer itself, consistent with the observations of Frehser.[15] In any case it is clear that differences in expansion characteristics of adjacent zones in the forged steel have a profound effect on the distortion of machined components in heat treatment.

TABLE 5 *Distortion measurements of annealed and hardened specimens*

Quench	Specimen				
	1	2	3	4	5
Furnace cooled	−1·6	−4·0	−3·0	+2·4	− 3·2
Oil	+2·9	+9·2	+3·4	−4·5	+10·0
„	+5·5	+6·2	+7·3	−5·1	+ 7·7
„	+4·9	+5·7	+6·8	−6·9	+ 2·8
„	+6·1	+4·6	+5·9	−6·4	+ 4·0

The difference in dilatation between the outermost and the intermediate zone might be expected to affect the distortion of angle sections machined entirely from those zones. In fact, the maximum deflection of specimen 2 is greater than that of specimen 1 after most heat treatments, but statistical tests on the rather small number of experiments do not establish significance.

These results establish the possibility of warping



28 *Method of carrying out distortion measurement*

of die steels due to differential anisotropy of dilatations in adjacent layers differing in carbide structure.

PRACTICAL APPLICATIONS OF THE DISTORTION TEST

The proof that distortion could be produced by stresses between adjacent layers differing in carbide structure and expansion characteristics was established by an experiment which could form a sound basis for a practical distortion test. A test of this nature has been employed to study the behaviour of specimens consisting of adjacent portions with different structures, in reasonable simulation of dies machined from forged bars.[14, 18, 20] The investigation covered the response of such specimens to horizontal and vertical quenching, to uneven heating for hardening, and to martempering; it was hoped to distinguish the effects of these special treatments from the influence of the carbide structure and to consider their mutual interaction. The study extended to an examination of various possibilities of achieving a more uniform carbide distribution, by homogenizing treatments and by modifications—bearing in mind practical limitations —of forging techniques.

The test described in the previous section was supplemented by a very similar one[14] employing a specimen characterized by a change in cross-section half-way along its length, which was geometrically more susceptible to distortion due to cooling and transformation stresses. The specimens were cut from other lengths of the same forged bar. The carbide distribution was assumed to be the same and this was subsequently confirmed by macro- and micro-examination of the distortion specimens after completion of the tests. Deviations from this distribution in other forged bars was detected in surveys of the carbide structure.

The shape of the specimens and the way they were cut from the forged bar, are shown in fig. 27. Specimen *B* was taken from the core of the billet characterized by a very coarse and somewhat random carbide structure. Specimens *A* and *C* came from the outside, where the longitudinal alignment of the carbides is more sharply pronounced and where there is an appreciable refinement of the carbides in the outermost zone. A fourth specimen, marked *D*, was taken from some of the billets. It is located near the surface of the billet but cut in such a way that the direction of bending due to the geometry of the specimen is at right angles to any bending introduced by differences in carbide distribution. In general specimen *D* exhibited relatively slight distortion.

One half of each specimen was $\frac{3}{4}$ in. square and in the case of specimens *A* and *C* this involved the presence of adjacent layers which differed in the degree of hot working and in carbide structure. The other half was effectively a continuation of one of these layers, the outermost one in specimen *C*, the intermediate one in specimen *A*. At the extreme end of the specimen the full square section was retained; this was not expected to influence the cooling rate but greatly facilitated grinding of the specimens on a magnetic plate.

For distortion measurements the thick end of the specimen was supported on a 4-in. ground parallel on a marking table and the position of the overhanging thin end was measured with a dial-gauge supported on a Vernier height gauge with its base on the same marking table. The method is illustrated in fig. 28. Dial-gauge readings were taken at points *a* and *b*; the mean was taken as the datum line for distortion. The deviation, in thousandths of an inch, of the thin end of the specimen from the datum line, reading *c* or *c'*, was recorded as the index of distortion—*c* corresponded to positive distortion, *c'* to negative. The measurements were carried out along both sides of the top face and the distortion figures averaged to reduce errors due to surface oxide, etc.

to be continued

BOOKS

A symposium on aluminium in building

The Aluminium Development Association, 1960. Pp. 256. £2 10s.

PROCEEDINGS of a symposium on aluminium in building at the Royal Institute of British Architects, London, on July 9 and 10, 1959, are given. The publication comprises the final version of 13 papers, presented at three sessions, together with the discussion which concluded each session.

The Association convened this symposium in order to provide opportunity for architects, builders and engineers on the one hand to exchange ideas and information with the aluminium industry on the other. The event was timely in view of the developments of interest to the building industry, coming largely from the aluminium industry and its allied trades over the past few years, including larger extruded sections and longer and wider sheets; improvements in joining and finishing techniques, and researches directed toward greater efficiency in design.

The book is very well illustrated, partly in colour, and falls naturally into three divisions: Aluminium as a building material (three papers), Aluminium curtain walling (four papers), and Aluminium building applications (six papers). The second session included surveys of curtain walling in this country, in North America, and in Europe, and the first of these papers gives section and assembly drawings of 16 British systems with photographs of typical installations. Other papers deal with economic aspects; standards and codes of practice; finishes; roofing and vertical cladding; windows; furniture; and with prefabricated and transportable buildings; also, in two papers, with research into problems of thermal insulation, and aluminium in relation to the Thermal Insulation Act. Discussion ranges from the uses of foil to construction of big domes, from the fire hazard to the supply of aluminium nails, from mastics to maintenance.

Industrial electric furnaces and appliances

By V. Paschkis and J. Persson. Interscience Publishers Inc., New York, 1960. Second edition. £9 net.

MOST METALLURGISTS and those who are responsible for their training, treat furnace design in an offhand and qualitative manner. They feel that the subject is so specialized and esoteric that there are no simple principles of furnace design—experience is what counts. However, in recent years the principles of heat transfer have been applied increasingly to metallurgical operations and the present volume reviews the present status of all classes of electric furnaces, their development and the theory upon which it is based. A special point is made of the relation of furnace design to the uniformity of the product to be heated. The work is well documented with many references.

The first chapter, occupying about one-fifth of the book, deals with the fundamentals of heat transfer and the application of Hessler or Guy-Laurie charts to calculate heating cycles for slabs, cylinders and spheres. The important principles are clearly explained with worked examples and without recourse to advanced mathematics. Steady heat flow principles are explained and used to calculate heat losses and wall thickness of furnaces.

In subsequent chapters, all the industrial electrical furnace types and heating methods are subjected to a detailed survey and design data are reviewed for an impressive list of furnace types including open-arc melting furnaces, submerged arc smelting furnaces, resistance heated furnaces, electrode salt bath furnaces, low temperature ovens, induction melting furnaces of all types and induction heating and dielectric heating appliances. For each type, the materials of construction, the operating conditions and characteristics are presented in detail with the aim of selecting the conditions of metallurgical as well as thermal efficiency. Finally, the factors affecting the choice of furnace types for a given process in which several down-to-earth observations are made concerning furnace efficiency and operation. These few pages at least should be made compulsory reading for every young metallurgist who aspires his keep by seeing that pieces of metal are heated and cooled at the correct rates with the hope that his employers will obtain a reasonable return on his activities.

This book is a unique blend of the theory and practice of electric furnace design and should find a place on the shelf of all furnace builders who aspire to be furnace designers and of all metallurgists who want to use furnace equipment to the best advantage. Manufacturers of gas- and oil-fired furnaces will also profit from this book, for as the author observes, the selection of furnaces is based often on incomplete consideration of fundamental questions concerning the selection of process, unit size and of principles of operation.

Here is a useful book of reference for most metallurgists and engineers. It is hardly suitable as a textbook, but perhaps its appearance will help to stimulate interest in the neglected subject of rational furnace design amongst teachers of industrial metallurgy.

A. D. H.

Vacuum processing in metallurgy

By J. Wesley Cable. Reinhold Publishing Corp., New York; Chapman & Hall, London, 1960. Pp. 202. £2 4s. net.

THE AIM of this American book is to present a record of the application of vacuum technology in metal treatment. Firstly, the physical aspect of the subject is discussed in a simple and qualitative manner. Methods of producing low pressures, their measurement and leak detection are similarly described, but due attention is paid to the selection of vacuum equipment for a particular application.

There follows a series of chapters dealing with the use of low pressure treatments in outgassing, refining distillation, melting and casting, vacuum deposition, powder metallurgy, brazing and welding and heat treatment. The furnaces, pumping equipment and processing techniques are described concisely from the point of view of the engineer. Metallurgical aspects are given less attention but are not overlooked.

The book is readable and succinct and gives an up-to-date review of American vacuum practice. Any metallurgist or engineer who is required to interest himself in the industrial applications of vacuum technology will find that this book gives a readable introduction to the subject. A serious omission is a lack of references for further detailed study.

A. D. H.

Overseas newspapers and periodicals guide book

Publishing & Distributing Co. Ltd., London, 1960. Seventh edition. £1 10s. net.

THIS WELL-KNOWN ANNUAL is a reference book covering markets all over the world and all trades from agriculture to zinc manufacture. It has been considerably enlarged in this new edition and gives details of over 3,000 publications in 137 foreign countries. Frequencies, advertising rates, subscription rates and circulations are indicated.

INSTRUMENTATION

Pure dry hydrogen

Deoxo catalytic purifiers have become widely accepted as the most efficient and economical method of removing oxygen from gases. One very important use lies in the purification of hydrogen as an atmosphere for heat treatment and other applications.

To meet the demand for a self-contained unit, combining the features of the Deoxo purifier with those of an automatically operated drying unit containing Linde molecular sieve, Engelhard Industries Ltd. have produced the Deoxo Puridryer.

This unit, having a maximum capacity of 100 s.c.f.h., will purify normal commercial-grade hydrogen to give a gas with an oxygen content of less than 1 part/million. By use of Linde molecular sieve as the drying agent it achieves a dew point as low as −100° F. (−73° C.) which enables the gas to be fed directly to a furnace.

The platinum metals catalyst used in the Puridryer to remove oxygen requires no re-activation, and unless poisoned by sulphur compounds, carbon monoxide, chlorine, and some organic compounds—which are rarely found in commercial hydrogen—should last indefinitely.

The Puridryer may also be used with other gases such as nitrogen, argon, neon and saturated hydrocarbons, providing that two volumes of hydrogen are present in the gas to react with each volume of oxygen impurity to be removed.

Automatic crack detection

An automatic universal magnetic crack detector has been developed by Fel-Electric Ltd. for the automatic indexing, simultaneous magnetization and inking of repetition ferrous castings and forgings of 3 to 12 inches long. The LB.167 houses all the equipment required to carry out automatically all the functions hitherto required of a magnetic crack detector operator so that the operator is called upon only to load and inspect the parts tested.

An eight-spoked casting is machined to carry suitable support cradles in which the components are allowed to rest. By a clockwise working indexing mechanism, each component is carried to the testing position where the movement is positively arrested. Gripping pressure is then applied to the ends of the component through resilient contact face-plates, after which the energizing fields are applied to magnetize the component for revealing surface flaws and defects in all directions.

1 *Universal magnetic crack detector*

The magnetic crack-detecting ink is applied in the form of a fine spray to all surfaces of the component during the magnetizing period. After inking and energizing a component passes out of the testing position by a repetition of the indexing process. The period elapsing between the component leaving the testing station and arriving back to the inspector allows all surplus ink to drain from the part, so giving an improved indication of flaw or defect.

Electronic grinding indicator

B. O. Morris Ltd. have been producing this instrument for several years for use in Rolls-Royce factories, and have now released it for general use. The purpose of the instrument is to show visually when a grinding wheel is in contact with the work, when the wheel itself is not visible due to being inside the work, inaccessible or obscured by coolant (fig. 2).

A pick-up head, mounted on a magnet, is positioned near the grinding wheel, and the vibrations therefrom produce signals in the pick up, which the instrument shows as a 'vee' on the cathode ray tube. Different kinds of 'vee' signals enable the operator to see when the wheel makes contact with the work, if he is grinding too heavily, and even reveal if eccentricity is present. Being portable the instrument is versatile and can be moved from job to job on different machines.

2 *Grinding indicator*

While the apparatus is normally supplied for 200/250 volts single phase, 50 cycles a.c., other a.c. voltages can be accommodated through transformer tappings. Full details are shown in Catalogue ME.1 from B. O. Morris Ltd., Morrisflex Works, Briton Road, Coventry.

Betatron for hot state radiography

Pantak Ltd., of Vale Road, Windsor, Berks., have concluded an agreement with Sierex Ltd. for the marketing and servicing of the Siemens 15-meV. industrial betatron. The introduction of this equipment expands Pantak's range of industrial X-ray equipment by adding a fully mobile unit which is capable of examining steel thicknesses of up to 16 in.

An important aspect of the use of the betatron is the use of an enlargement technique which gives high flaw sensitivity. Enlargements of up to four times normal size are possible and enables the detection of fine flaws which would otherwise be difficult or impossible to observe. A further significant development is the application of hot-state radiography to welds or billets. This opens up new possibilities for establishing production techniques, with the elimination of costly defects.

NEWS

British industrial consultants' link with Common Market

To develop trading opportunities with the European Common Market and to encourage closer exchange of industrial and management 'know-how' between Britain and Europe, Production-Engineering Ltd. recently founded a new organization in Paris.

The new organization, Production-Engineering Société Anonyme, 3 Square de l'Opera, Paris 9e, a member of the PE Group of companies, has already carried out several assignments in France, including an important one for the French ship-building industry. Prominent among its specialized services is electronic data processing.

British Oxygen Co.

The British Oxygen Co. Ltd. has resumed trading under its own name. The following companies continued as agents for the parent company until December 31, 1960, after which date they ceased to operate: British Oxygen Gases Ltd., British Oxygen Engineering Ltd., British Oxygen Research & Development Ltd., Quasi-Arc Ltd., Sparklets Ltd., and A. Charles King Ltd.

All current trade marks and trade names will continue to be used. This means that, whilst British Oxygen Gases Ltd. and Quasi-Arc Ltd. as such cease to operate, all the well-known British Oxygen and Quasi-Arc and allied trade marks will continue to be used for products in this field. A new Sales Division of British Oxygen will be responsible for all group selling activities in the U.K., excluding chemicals.

A new Technical Division is to be set up. It is proposed to enlarge the technical centre at Cricklewood which will have enhanced importance through concentration of technical development of products and processes backing up the whole of the home and export selling effort.

LECTURES

Modern developments in the theory and practice of steel heat treatment

Eight lectures at the College of Advanced Technology, Gosta Green, Birmingham, at 6.30 p.m., on successive Tuesday evenings, beginning January 17, 1961. *Course Tutor:* A. D. Hopkins, M.SC., A.I.M. Each meeting will last for about two hours and will include a lecture, followed by a discussion.

January 17. Some structural and physical aspects of the heat treatment of steels. Dr. K. W. Andrews.

January 24. Quenching and quenching media. Dr. R. T. Allsop.

January 31. The avoidance of distortion in heat treatment. Dr. K. Sachs.

February 7. Isothermal heat treatment. Mr. J. C. Gregory.

February 14. Modern tool steels. Mr. E. Johnson.

February 21. The treatment, properties and application of high-strength steels. Dr. C. L. M. Cottrell.

February 28. Fuel efficiency in heat treatment. Mr. G. A. Blackwell.

March 7. Costing of heat-treatment operations. Mr. E. P. Wilson.

Metallurgy of deep drawing and pressing

Ten lectures at the College of Advanced Technology, Gosta Green, Birmingham, at 6.30 p.m., on successive Wednesday evenings, beginning January 18, 1961. *Course Tutor:* J. C. Wright, B.SC., PH.D., A.I.M., A.M.INST.F., A.C.T.BIRM. Each meeting will last for about two hours and will include a lecture, followed by a discussion. Demonstrations will be arranged in the laboratories on a variety of detailed testing procedures. *Demonstrator:* L. W. Crane, B.SC.

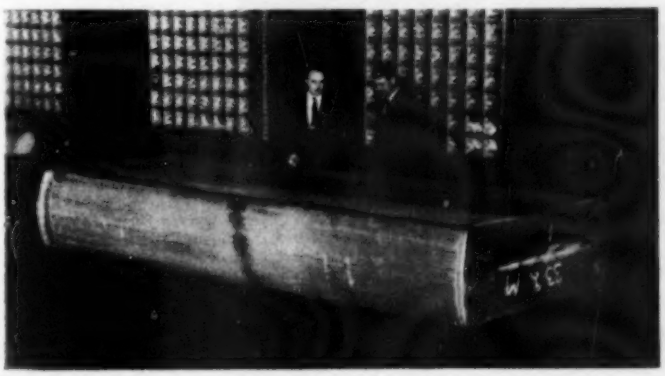
UNIVERSITY OF LONDON

The electronic structure of metals

A course of three lectures by Professor A. B. Pippard, M.A., PH.D., F.R.S., will be given at Imperial College of Science and Technology, New Physics Building, Queen's Gate, S.W.7, at 5.30 p.m. on Thursdays, February 2, 9 and 16. The lectures are addressed to students of the University and to others interested in the subject. Admission free, without ticket.

The theory of the alloys of copper

Professor W. Hume-Rothery, O.B.E., F.R.S., of Oxford University, is to give a lecture on 'The theory of the alloys of copper' under the auspices of the Metal Physics Committee of the Institute of Metals. The meeting will take place at the Royal Institution, Albemarle Street, London, W.1, on Thursday, February 23, 1961, at 6.30 p.m. Visitors are welcome; tickets are not required.



THE LARGEST *rectangular aluminium ingot yet made has been cast by Reynolds Metals Co.'s sheet and plate plant in the U.S.A. for use in the company's programme to build the Aluminaut—the aluminium research submarine designed to explore the ocean's depths three miles down. The* 31,750-*lb. ingot, measuring* 26 × 72 × 180 *in., will be forged into the first of several cylindrical rings which, when joined together, will form the submarine's* 48-*ft. long hull. Aluminium is believed to be the only practical material strong enough to resist the external pressures at superdepths. A similar-sized steel hull would be too heavy to return to the surface.*

PEOPLE

MR. C. R. WHEELER, C.B.E. (chairman, Guest Keen Iron and Steel Co. Ltd., additional vice-chairman, Associated Electrical Industries Ltd., and a director of the Steel Company of Wales Ltd.), succeeded Mr. R. F. Summers (chairman, John Summers & Sons Ltd.) as president of the British Iron and Steel Federation on the 1st of this month.

Mr. C. R. Wheeler entered the steel industry on leaving school in 1922, when he joined Baldwins Ltd. After being London manager for the Port Talbot Steel Co. and other interests of Baldwins Ltd., he joined Guest, Keen, Baldwins Iron and Steel Co. Ltd. on its formation in 1930, later becoming commercial manager and, in 1940, a director.

During the war Mr. Wheeler served in the Iron and Steel Control of the Ministry of Supply, at first as assistant controller, Raw Materials. In 1943 he became deputy controller, Raw Materials, and in that year headed the British delegation to the Joint Metallurgical Committee in the U.S.A., which considered and reported on alloy steel and alloy conservation. In 1945 he was appointed controller of Iron and Steel. For his war-time services he was awarded a C.B.E. in 1946.

In May, 1946, Mr. Wheeler returned to Guest, Keen, Baldwins Ltd. (later Guest Keen Iron and Steel Co. Ltd.) as joint managing director. In December, 1959, he was appointed chairman of Guest Keen Iron and Steel Co. Ltd.

Mr. Wheeler has recently been appointed an additional vice-chairman of Associated Electrical Industries Ltd. and is to take up executive duties with that company in 1962. He has therefore resigned from the chairmanship of Guest Keen Iron and Steel Co. Ltd. at the end of last year. He remains a director of Guest Keen and Nettlefolds Ltd., however, and has recently been appointed a director of the Steel Company of Wales Ltd.

Mr. Wheeler has been a member of the Council and the Executive Committee of the British Iron and Steel Federation since 1946. As chairman of BISC (Ore) Ltd. since its formation in 1946, he has been directly concerned with the expansion of the industry's central ore importing arrangements, including the development of the ore carrier fleet and investment in overseas orefields. He has also been chairman of the Federation's International Trade Relations Committee since 1958 and a director of the industry's main central importing agency, BISC Ltd.

Mr. Wheeler has represented the Government on the Steel Committee of the Economic Commission for Europe and the Committee of European Economic Co-operation and was present as an adviser at the Foreign Ministers' Conference in Moscow in 1947.

He was High Sheriff of the County of Glamorgan in 1955 and president of the Iron and Steel Institute in 1958-59. He is a Chevalier de la Legion d'Honneur and a member of the Court of Governors of the University College of South Wales and Monmouthshire. Mr. Wheeler was born in 1904 and educated at St. Paul's School. He is married and has one son and two daughters.

Sir Julian Pode has been named as president-elect of the British Iron and Steel Federation for 1961.

Sir Julian Pode is managing director of the Steel Company of Wales Ltd. He entered the steel industry in 1926, when he joined the staff of Guest, Keen and Nettlefolds Ltd. as district accountant at the Dowlais works, and in 1930, when the company's heavy steel interests were amalgamated with those of Baldwins Ltd., he was appointed secretary of the new company, Guest, Keen, Baldwins Iron and Steel Co. Ltd. In 1938 he became secretary and joint commercial manager and in 1943 was appointed assistant managing director and in 1945 joint managing director. On the formation of the Steel Company of Wales Ltd. in September, 1947, he was appointed managing director of the new company.

Sir Julian Pode has been a member of the Council and the Executive Committee of the British Iron and Steel Federation since 1945, is chairman of its Price Policy Committee and serves on a number of other Federation committees. He is honorary treasurer of the Iron and Steel Institute, a member of the Industrial Coal Consumers' Council, a director of the National Industrial Fuel Efficiency Service, a member of the Grand Council of the Federation of British Industries and chairman of its Fuel and Power Consumers' Policy Committee.

He is deputy chairman of the Development Corporation of Wales, a director of Guest, Keen, Nettlefolds Ltd., the Steetley Co. Ltd. and other public companies.

Sir Julian was High Sheriff of Glamorgan in 1948 and was appointed a Justice of the Peace in 1951. He was made an Honorary Freeman of the Borough of Port Talbot in March, 1957. A knighthood was conferred upon him in the New Year Honours of 1959. He was born in 1902, was educated in H.M.S. *Conway* and subsequently qualified as a chartered accountant. He served in the Royal Navy during the first world war. He is married and has a son and a daughter.

British Insulated Callender's Cables Ltd. announces the retirement of **Mr. T. E. H. Birley,** O.B.E., deputy overseas manager, after 40 years' service with the company.

Mr. Birley was born in Manchester in 1898 and

Mr. T. E. H. Birley, O.B.E.

educated at Marlborough and Pembroke College, Cambridge, where he won a Classical Scholarship in 1916. He served as a pilot in the Royal Flying Corps, being wounded and taken prisoner in 1918. In 1920 he joined Macintosh Cable Co. Ltd. He transferred to British Insulated Cables Ltd. in 1936, when Macintosh became absorbed by the B.I. Co. and served that company on many trade association committees until the outbreak of war.

From 1942 to 1945 Group Captain Birley, as he was then, was assistant secretary on the War Cabinet Secretariat in Washington and for his services there he received in 1947 the American Legion of Merit—Degree of Officer. The importance of his work in America was also recognized in the 1946 New Year Honours List by the award of the O.B.E.

After the war he returned to BICC as special representative on Overseas Sales and in 1950 became regional

manager for Central and South America. In 1957 he was appointed deputy overseas manager.

Mr. A. R. O. Williams, O.B.E., A.R.S.M., B.SC., a managing director of Consolidated Gold Fields of South Africa Ltd., and of New Consolidated Gold Fields Ltd., has been elected president of the Institution of Mining and Metallurgy for 1961-62.

He was educated at Malvern and trained in metallurgy at the Royal School of Mines from 1922 to 1926, and after a brief period in Magdeburg he went to South Africa as a learner miner at Robinson Deep, Johannesburg, beginning an association with New Consolidated Gold Fields unbroken except for a few short periods and war service.

Mr. Williams gained further experience at Kansanshi mine of Rhodesia-Katanga Co. Ltd., Northern Rhodesia, from 1930 to 1933, and then returned to England on his appointment as assistant to the chief engineer at the London head office of New Consolidated Gold Fields. In 1938 his services were lent to Gold Fields American Development Co. Ltd., and he took up residence in Toronto as their engineer and representative in Canada.

During the war years Mr. Williams served with distinction in the Royal Engineers. He was lieutenant-colonel commanding military tunnelling operations at Gibraltar during 1942-43, and later served in France, Belgium, Holland and Germany. He was awarded the O.B.E. in 1944 and mentioned in despatches.

In 1945 Mr. Williams rejoined New Consolidated Gold Fields Ltd. as resident engineer in London. His work has entailed visits to mines in many parts of the world. Apart from the Gold Fields Group, he has been a member of the board of directors of various mining companies.

Mr. Williams joined the Institution as a student in 1926 and was elected an associate member in 1932 and a member in 1944. His service on the Council covers 15 years, from 1946 to the present time, and includes the office of vice-president during the period 1951-54. He will take office in May 1961 in succession to Professor David Williams, D.SC., PH.D.

Under a suggestions scheme run by the Morgan Crucible Co. Ltd., **Mr. James H. Fairweather,** laboratory assistant in the company's Research Department, has been awarded £225. His alteration to the cooling system of a laboratory furnace will save approximately 10 million gallons of water a year, and represents an annual saving to the company of £400.

Mr. Fairweather's success comes after 21 years with the company, which already this year has paid out £790 to employees under the scheme.

Mr. F. Baillie has joined the Cambridge Instrument Co. Ltd. as production manager and will be responsible for all aspects of production planning at the company's three factories in the United Kingdom. Mr. Baillie was formerly general works manager at the Greenock factory of I.B.M. (United Kingdom) Ltd.

During the war Mr. Baillie worked for a short time on radar equipment at the Air Ministry and later joined the Royal Air Force, where he did similar work as a flight-lieutenant in the signals branch.

Mr. R. Lewis Stubbs, director of the Zinc Development Association, has been appointed director-general of the Lead Development Association and the Zinc Development Association. Mr. Stubbs will be responsible for the overall direction and expansion of the work of both organizations and will represent them at international meetings. Each association will continue to function separately and general managers for both will be appointed.

Samuel Fox & Co. Ltd., a subsidiary of the United Steel Companies Ltd., announces that **Mr. R. D. Pollard,** at present director and chief metallurgist, has been appointed director of metallurgy. **Mr. R. Willcock,** at present works metallurgist, becomes chief metallurgist, responsible to the director of metallurgy for the management of the company's metallurgical department.

Mr. J. C. Colquhoun has retired as chairman of the Manganese Bronze and Brass Co. Ltd. **Mr. J. C. Budd** has been appointed chairman and **Mr. R. Dennis Poore** has been appointed a director.

Mr. James Clifton Colquhoun was born 67 years ago in a copper mining camp in Arizona. His father was president of the Arizona Copper Co., now owned by the Phelps Dodge Corporation. Educated at Cambridge, Mr. Colquhoun served in the first world war, during which he was awarded the M.B.E. and was Mentioned in Dispatches.

In 1932, at the age of 39, Mr. Colquhoun became chairman of the Manganese Bronze and Brass Co. Ltd. For some years he combined with his position as chairman that of managing director, and from 1945-59 he was also chairman of Lightalloys Ltd.

He has served on the Grand Council of the Federation of British Industries, as hon. treasurer of the Institute of Metals and has been a member of the Councils of the British Non-Ferrous Metals Federation, the British Non-Ferrous Metals Research Association and the British Bureau of Non-Ferrous Metals Statistics.

Mr. A. Broomhead has been appointed deputy managing director of Thos. Firth and John Brown Ltd. and **Mr. F. Wortley,** a director of Thos. Firth and John Brown Ltd., has been appointed general works manager.

PEOPLE

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Sir Julian Pode has been named as president-elect of the British Iron and Steel Federation for 1961.

Sir Julian Pode is managing director of the Steel Company of Wales Ltd. He entered the steel industry in 1926, when he joined the staff of Guest, Keen and Nettlefolds Ltd. as district accountant at the Dowlais works, and in 1930, when the company's heavy steel interests were amalgamated with those of Baldwins Ltd., he was appointed secretary of the new company, Guest, Keen, Baldwins Iron and Steel Co. Ltd. In 1938 he became secretary and joint commercial manager and in 1943 was appointed assistant managing director and in 1945 joint managing director. On the formation of the Steel Company of Wales Ltd. in September, 1947, he was appointed managing director of the new company.

Sir Julian Pode has been a member of the Council and the Executive Committee of the British Iron and Steel Federation since 1945, is chairman of its Price Policy Committee and serves on a number of other Federation committees. He is honorary treasurer of the Iron and Steel Institute, a member of the Industrial Coal Consumers' Council, a director of the National Industrial Fuel Efficiency Service, a member of the Grand Council of the Federation of British Industries and chairman of its Fuel and Power Consumers' Policy Committee.

He is deputy chairman of the Development Corporation of Wales, a director of Guest, Keen, Nettlefolds Ltd., the Steetley Co. Ltd. and other public companies.

Sir Julian was High Sheriff of Glamorgan in 1948 and was appointed a Justice of the Peace in 1951. He was made an Honorary Freeman of the Borough of Port Talbot in March, 1957. A knighthood was conferred upon him in the New Year Honours of 1959. He was born in 1902, was educated in H.M.S. *Conway* and subsequently qualified as a chartered accountant. He served in the Royal Navy during the first world war. He is married and has a son and a daughter.

British Insulated Callender's Cables Ltd. announces the retirement of **Mr. T. E. H. Birley**, O.B.E., deputy overseas manager, after 40 years' service with the company.

Mr. Birley was born in Manchester in 1898 and

Mr. T. E. H. Birley, O.B.E.

educated at Marlborough and Pembroke College, Cambridge, where he won a Classical Scholarship in 1916. He served as a pilot in the Royal Flying Corps, being wounded and taken prisoner in 1918. In 1920 he joined Macintosh Cable Co. Ltd. He transferred to British Insulated Cables Ltd. in 1936, when Macintosh became absorbed by the B.I. Co. and served that company on many trade association committees until the outbreak of war.

From 1942 to 1945 Group Captain Birley, as he was then, was assistant secretary on the War Cabinet Secretariat in Washington and for his services there he received in 1947 the American Legion of Merit—Degree of Officer. The importance of his work in America was also recognized in the 1946 New Year Honours List by the award of the O.B.E.

After the war he returned to BICC as special representative on Overseas Sales and in 1950 became regional

manager for Central and South America. In 1957 he was appointed deputy overseas manager.

Mr. A. R. O. Williams, O.B.E., A.R.S.M., B.SC., a managing director of Consolidated Gold Fields of South Africa Ltd., and of New Consolidated Gold Fields Ltd., has been elected president of the Institution of Mining and Metallurgy for 1961-62.

He was educated at Malvern and trained in metallurgy at the Royal School of Mines from 1922 to 1926, and after a brief period in Magdeburg he went to South Africa as a learner miner at Robinson Deep, Johannesburg, beginning an association with New Consolidated Gold Fields unbroken except for a few short periods and war service.

Mr. Williams gained further experience at Kansanshi mine of Rhodesia-Katanga Co. Ltd., Northern Rhodesia, from 1930 to 1933, and then returned to England on his appointment as assistant to the chief engineer at the London head office of New Consolidated Gold Fields. In 1938 his services were lent to Gold Fields American Development Co. Ltd., and he took up residence in Toronto as their engineer and representative in Canada.

During the war years Mr. Williams served with distinction in the Royal Engineers. He was lieutenant-colonel commanding military tunnelling operations at Gibraltar during 1942-43, and later served in France, Belgium, Holland and Germany. He was awarded the O.B.E. in 1944 and mentioned in despatches.

In 1945 Mr. Williams rejoined New Consolidated Gold Fields Ltd. as resident engineer in London. His work has entailed visits to mines in many parts of the world. Apart from the Gold Fields Group, he has been a member of the board of directors of various mining companies.

Mr. Williams joined the Institution as a student in 1926 and was elected an associate member in 1932 and a member in 1944. His service on the Council covers 15 years, from 1946 to the present time, and includes the office of vice-president during the period 1951-54. He will take office in May 1961 in succession to Professor David Williams, D.SC., PH.D.

Under a suggestions scheme run by the Morgan Crucible Co. Ltd., **Mr. James H. Fairweather,** laboratory assistant in the company's Research Department, has been awarded £225. His alteration to the cooling system of a laboratory furnace will save approximately 10 million gallons of water a year, and represents an annual saving to the company of £400.

Mr. Fairweather's success comes after 21 years with the company, which already this year has paid out £790 to employees under the scheme.

Mr. F. Baillie has joined the Cambridge Instrument Co. Ltd. as production manager and will be responsible for all aspects of production planning at the company's three factories in the United Kingdom. Mr. Baillie was formerly general works manager at the Greenock factory of I.B.M. (United Kingdom) Ltd.

During the war Mr. Baillie worked for a short time on radar equipment at the Air Ministry and later joined the Royal Air Force, where he did similar work as a flight-lieutenant in the signals branch.

Mr. R. Lewis Stubbs, director of the Zinc Development Association, has been appointed director-general of the Lead Development Association and the Zinc Development Association. Mr. Stubbs will be responsible for the overall direction and expansion of the work of both organizations and will represent them at international meetings. Each association will continue to function separately and general managers for both will be appointed.

Samuel Fox & Co. Ltd., a subsidiary of the United Steel Companies Ltd., announces that **Mr. R. D. Pollard,** at present director and chief metallurgist, has been appointed director of metallurgy. **Mr. R. Willcock,** at present works metallurgist, becomes chief metallurgist, responsible to the director of metallurgy for the management of the company's metallurgical department.

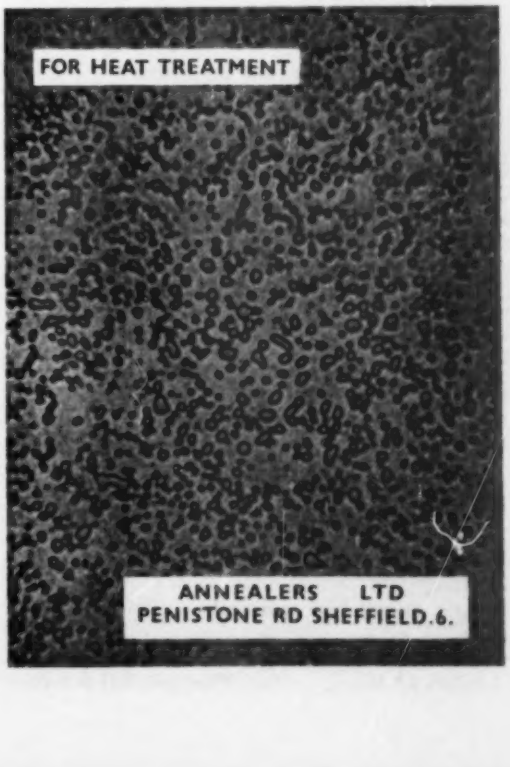
Mr. J. C. Colquhoun has retired as chairman of the Manganese Bronze and Brass Co. Ltd. **Mr. J. C. Budd** has been appointed chairman and **Mr. R. Dennis Poore** has been appointed a director.

Mr. James Clifton Colquhoun was born 67 years ago in a copper mining camp in Arizona. His father was president of the Arizona Copper Co., now owned by the Phelps Dodge Corporation. Educated at Cambridge, Mr. Colquhoun served in the first world war, during which he was awarded the M.B.E. and was Mentioned in Dispatches.

In 1932, at the age of 39, Mr. Colquhoun became chairman of the Manganese Bronze and Brass Co. Ltd. For some years he combined with his position as chairman that of managing director, and from 1945-59 he was also chairman of Lightalloys Ltd.

He has served on the Grand Council of the Federation of British Industries, as hon. treasurer of the Institute of Metals and has been a member of the Councils of the British Non-Ferrous Metals Federation, the British Non-Ferrous Metals Research Association and the British Bureau of Non-Ferrous Metals Statistics.

Mr. A. Broomhead has been appointed deputy managing director of Thos. Firth and John Brown Ltd. and **Mr. F. Wortley,** a director of Thos. Firth and John Brown Ltd., has been appointed general works manager.



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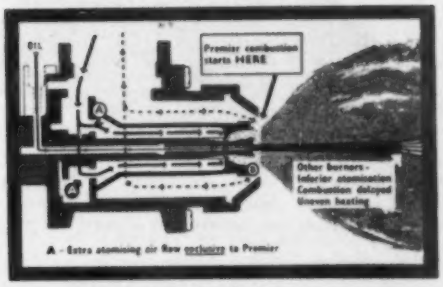
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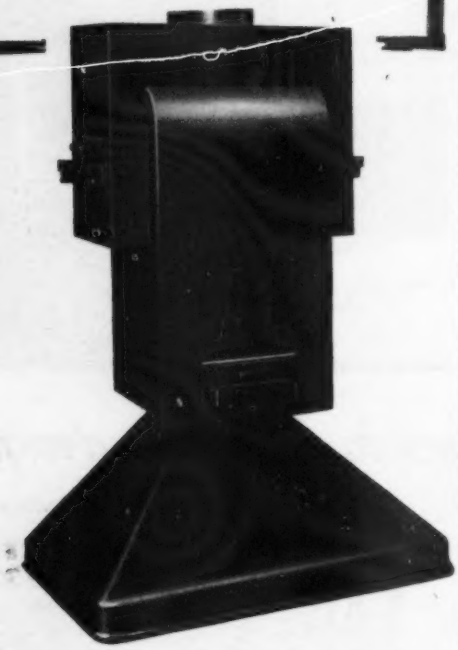

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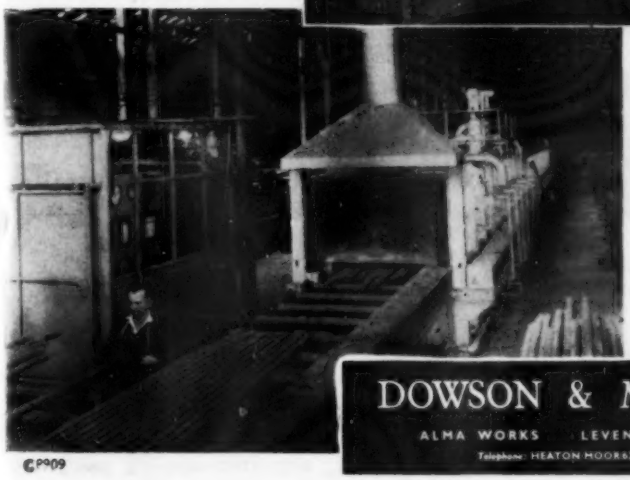

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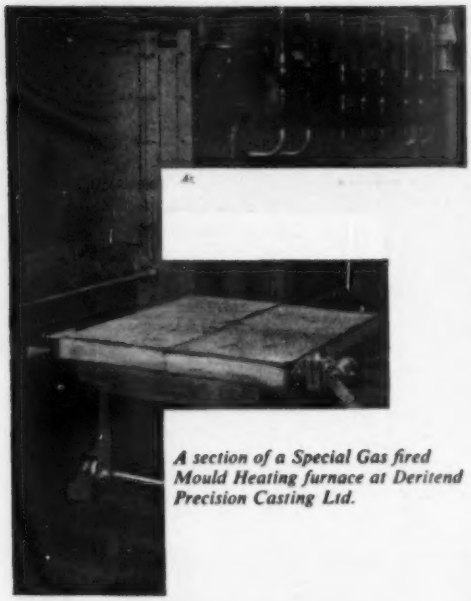

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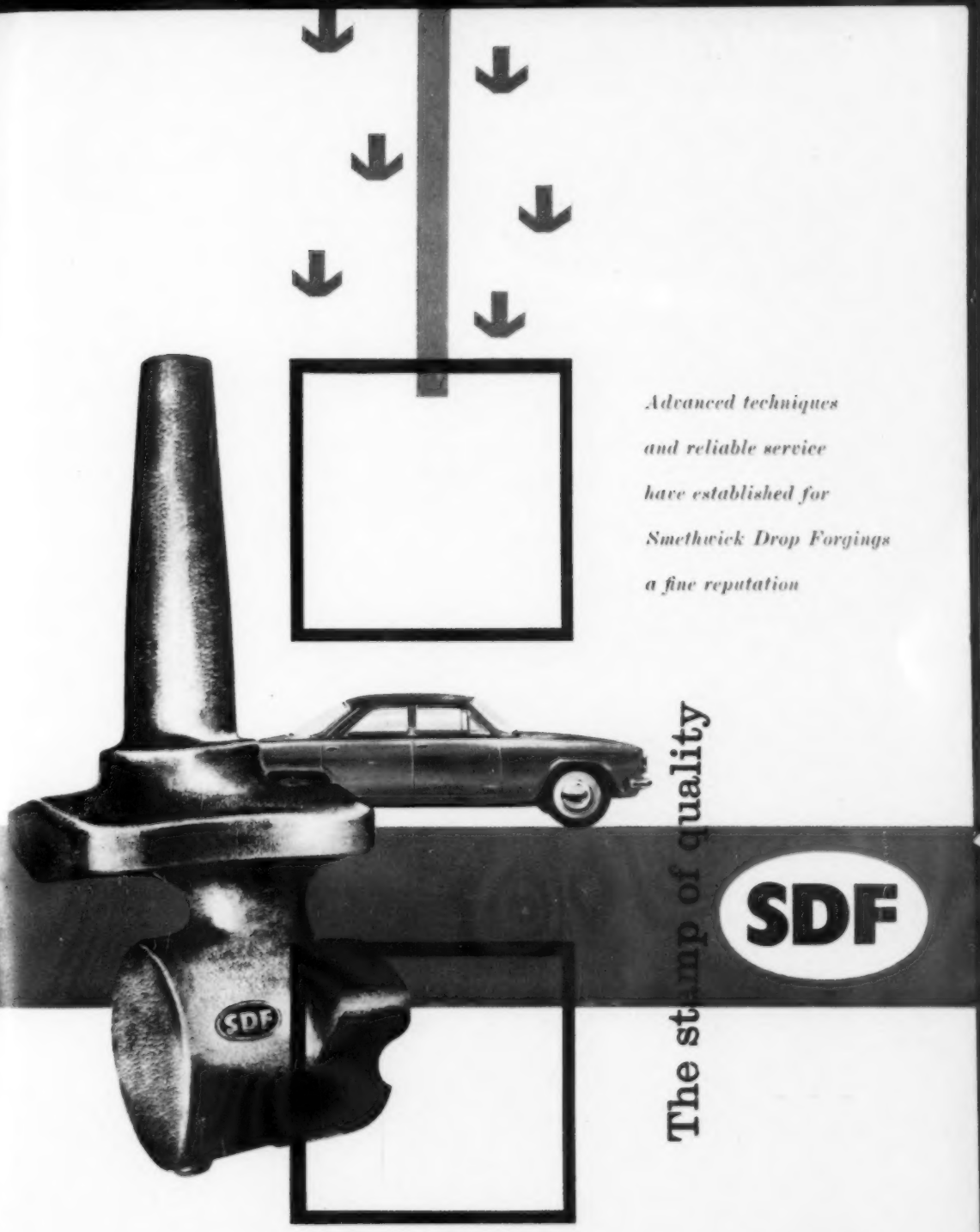

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